Metal Progress

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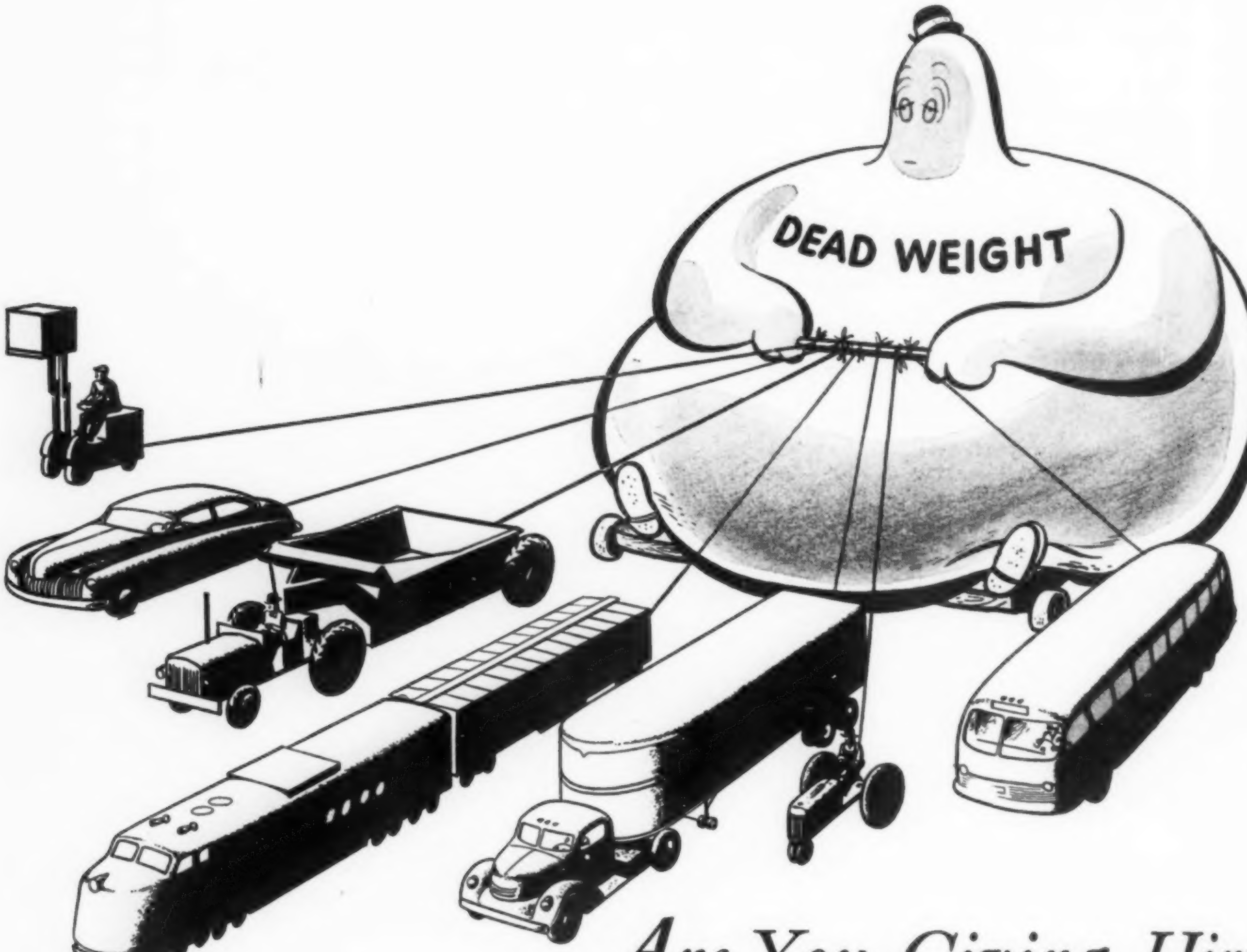
DEAD WEIGHT

Metal Progress

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Formation of the Natural Oxide Film on Aluminum

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SINCE the oxide films that form on aluminum and its alloys upon exposure to the atmosphere play a very important role in the performance of this metal, the behavior of these films is both of interest and importance. They are very adherent and — although generally so thin as to be invisible — they effectively limit further oxidation and make aluminum appear to have a greater passivity than would be expected for such an electronegative element. To a substantial extent, therefore, aluminum depends upon the protection afforded by the natural oxide film to resist oxidation and corrosion.

The unusual resistance of aluminum to oxidation was recognized by some of the early pioneers. For instance, H. Sainte-Claire Deville in 1859 stated in his book, "De L'Aluminium":

> "Aluminum is entirely unaffected by dry or moist air. This is one of the most valuable properties of the metal and one on which its practical use is based."

Deville, however, did not recognize that the resistance of aluminum to chemical attack was due to the ever-present natural oxide film. When he attempted to solder aluminum, he observed that this metal seemed to possess the inherent characteristics of resisting wetting by low-melting metals, such as might be used as solders. Actually, the existence of an oxide film on aluminum was first mentioned by Joseph W. Richards in 1890 in the second edition of his book, "Aluminium", in the following statement:

> ". . . . It is now well-known that objects of commercial aluminium do, after a long exposure, become coated with a very thin film which gives the surface a 'dead' appearance. . . . . . The oxidation, however, does not continue, for the film seems to be absolutely continuous and to protect the metal underneath from further oxidation."

At this time, the full significance of the protective action of the oxide film was not recognized. Professor Richards, however, was interested in the soldering of aluminum and encountered some difficulties. His remarks regarding the soldering of aluminum, published in 1896 in the third edition of "Aluminium", were as follows:

This is the first half of a paper prepared at the request of the Pittsburgh Chapter, ⚙, for presentation to the International Conference on Surface Reactions held in that city last month. It contains the results of some new experiments that attempt to solve a very difficult problem — to measure the rate of formation of a very thin, invisible film that builds up on clean aluminum at a very rapid initial rate. This well-rounded review will enable the reader to interpret published information on the subject and interested experimenters to plan future work.

"It is difficult to expose a bare surface of aluminium to the action of the solder. Grease and dirt can be easily enough removed, but there is always present a thin, continuous coating of alumina which effectively prevents the solder from getting to the metal underneath. It might be thought that a thorough sandpapering or filing of the surfaces immediately before soldering would overcome it, but we find it of absolutely no use; the reason is that the thin invisible skin of alumina forms instantly as quickly as the metal is laid bare."

Here we have clear-cut recognition of the natural oxide film on aluminum and of its important characteristic of re-forming or healing quickly, once it is broken. Since the natural oxide film that forms when aluminum is exposed to the atmosphere is so thin and transparent as to be invisible, the evidence for its existence was of an indirect nature — such as, for example, its action in connection with the soldering of aluminum.

Additional evidence of the indirect type of the presence of a protective oxide film on aluminum is given by the behavior of aluminum with mercury. For instance, a piece of commercial aluminum sheet can be immersed in mercury for an extended period without any action taking place. However, when a sample is submerged and its surface is scratched or abraded while under the mercury, amalgamation occurs immediately. Should the sample be scratched while exposed to the air and then immersed in mercury, no amalgamation will occur. Apparently a thin but protective oxide film re-forms very quickly, and this film is of sufficient thickness to prevent the mercury from wetting the aluminum and amalgamating with it.

Thus, our knowledge of oxide films on aluminum has increased from indirect evidence to qualitative information. With the modern development of X-ray and electron diffraction, as well as other precise experimental methods, it has been possible to obtain some quantitative data on the rate of formation, the structure, and the composition of oxide films on aluminum, which have been helpful in indicating the manner in which they form and behave in service.

Measurement of Rate of Oxidation

Measurement of the rate at which oxide films form on aluminum, as well as on other metals, has been the subject of great interest not only to those specializing in a study of surface reactions, but also to those concerned with the practical aspects of protecting metals against corrosion. Aluminum has held a unique interest in this field because it differs in behavior from many of the other common metals; moreover, its behavior is such as to present uncommon experimental difficulties. First, there is the problem of securing an aluminum surface free from oxide, adsorbed gases or other impurities. Second, determining the rate of oxidation by weighing is a difficult problem because, on a sample of a size practical for weighing, weight changes on the order of 0.5×10^{-6} g. must be measured in order to secure reasonable accuracy. Third, the rate of initial oxidation is very rapid and weighings must be made quickly. There are also other problems which will be discussed in this paper. In presenting this subject, some of the published work will be reviewed, both to point out the difficulties involved and to record the progress made. New experimental work in this field will also be described with the same objectives in view.

One of the first attempts to determine the rate of formation of the oxide film on aluminum at ordinary temperatures was made by Vernon.[1] In his experiments, samples of aluminum sheet were prepared for initial weighing by scratch brushing with a steel wire brush and were subsequently exposed to the atmosphere at room temperature in a basement of a building in London. They were weighed at various intervals and it was found that most of the increase in weight occurred during the first five days' exposure, although 10 to 14 days

[1] "Report to the Atmospheric Corrosion Research Committee", by W. H. J. Vernon. *Transactions* of the Faraday Society, V. 23, 1927, p. 150.

were required before the curve flattened out along the time axis. Vernon estimated the eventual film thickness to be of the order of 1×10^{-6} cm. or 100 A. The weight increases, however, were too small in comparison with the probable error of weighing to warrant much quantitative significance. The evidence was to the effect that the oxide film grows rapidly at first and that after the initial formation period it becomes substantially impervious and protective.

For investigating the rate of film formation on metals, Gulbransen[2] developed a sensitive quartz microbalance, operating in an all-glass vacuum system and giving weights with a reproducibility of about 0.3×10^{-6} g. Gulbransen and Wysong have published data on the weight and rate of film formation on thin aluminum sheet at temperatures from 350 to 475° C. (660 to 885° F.), and these experiments will be discussed later. Apparently they have not investigated the rate of formation at room temperature.

Rate of Oxidation in Dry Oxygen

Careful measurements of the rate of oxidation of aluminum at room temperatures have been made at Aluminum Research Laboratories by H. H. Podgurski. Weight increases during oxidation were determined on a specially constructed glass-enclosed microbalance with a quartz beam. This balance was similar in many respects to the one described by Gulbransen in *Metal Progress* for March 1946. The apparatus was arranged and the tests were conducted so that the balance and its case could be thoroughly baked and degassed before each experiment. Also the tungsten filament and the aluminum wire used for producing the thin film for the oxidation experiments were degassed above the melting point of aluminum. The aluminum surface to be oxidized was obtained by evaporating the aluminum *in vacuo* and condensing it as a film on a very thin quartz plate suspended from one arm of the balance. Great care was taken to avoid any diffusion of mercury or oil vapor from gas traps or vacuum pumps to the freshly evaporated aluminum films. During evaporation of the films, the system was continuously pumped out with a three-stage oil diffusion pump via a trap cooled in liquid nitrogen. The aluminum film on the quartz plate was stored in the evacuated balance case (pressure 10^{-6} mm. of mercury) until the start of the oxidation test. The sensitivity of the balance was such that observations could be made to about 0.5×10^{-6} g. The apparatus was so arranged that *dry* oxygen could be quickly introduced into the evacuated balance case, but this rush of oxygen into the system set up currents that interfered with steady balance readings for about 5 min. afterward.

The rates of oxidation of two aluminum films with this system are shown in Fig. 1. One film of condensed aluminum (A) weighed 7.0×10^{-6} g. per sq.cm., and the other, (B), weighed 11.7×10^{-6} g. per sq.cm. They were deposited on quartz plates 16 sq.cm. in area, so their average thickness would be 260 A and 430 A, respectively.

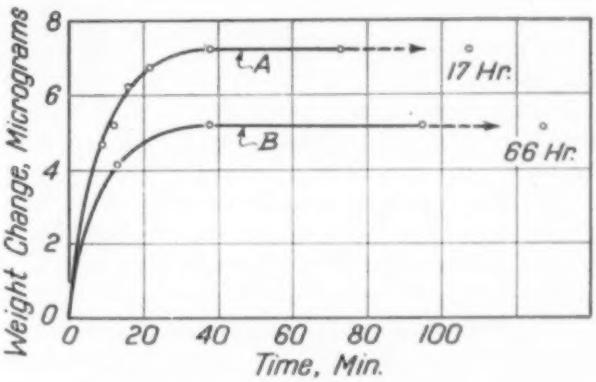


Fig. 1—Rate of Oxide Formation on Two Evaporated Aluminum Films, Exposed to Dry Oxygen at Room Temperature. Apparent surface area of each specimen was 16 sq.cm. and the oxygen pressure was 200 mm. of mercury. Sample A weighed 7.0×10^{-6} g. per sq.cm. and was about 260 Angstrom units thick; Sample B weighed 11.7×10^{-6} g. per sq.cm. and was about 430 Angstrom units thick. (*H. H. Podgurski*)

The most significant difference between the two curves shown in Fig. 1 is the final weight of the *oxide* film formed on the aluminum. The thinner film (A) of aluminum metal showed an increase in weight of 7.2×10^{-6} g., whereas the oxide film on the heavier aluminum deposit weighed only 5.2×10^{-6} g. In a third experiment, an aluminum film weighing 14×10^{-6} g. per sq.cm. acquired an oxide film weighing 5.2×10^{-6} g.

Apparently the initial rate of oxidation is rapid and a constant weight seems to be reached in a period somewhere between 10 and 40 min. In the case of the thinner metal deposit (A), there was no further increase in weight during exposure to oxygen up to 17 hr., and the heavier deposit (B) remained in oxygen for 66 hr. without any further change in weight.

As a check on these results, another experi-

[2] "A Vacuum Microbalance for the Study of Chemical Reactions on Metals", by E. A. Gulbransen; *Review of Scientific Instruments*, V. 15, 1944, p. 201. "Thin Oxide Films on Aluminum", by E. A. Gulbransen and W. S. Wysong; *Journal of Physical and Colloid Chemistry*, V. 51, 1947, p. 1087. "New Developments in the Study of Surface Chemistry", by E. A. Gulbransen; *Metal Progress*, V. 49, March 1946, p. 553.

ment was made in which an aluminum film of similar weight was exposed to dry nitrogen at a pressure of 200 mm. With this gas, a slight increase in weight was observed, but it amounted to only one microgram, which is about one fifth of the final weight of the oxide film that was formed in oxygen at the same pressure.

Mr. Podgurski attempted to establish the rate of film formation in air containing some water vapor. The results are not included, because of uncertainties arising from experimental difficulties. The observations indicated, however, that film growth under these conditions does not end as abruptly as it does in dry oxygen.

In attempting to interpret these data in terms of oxide film thickness, one must consider the fact that the condensed films of metallic aluminum, although they appear continuous to the eye, seldom are smooth and continuous when viewed under the microscope. Generally they also show a certain translucency to light, indicating the presence of minute pinholes. Furthermore, the quartz plate on which it deposits is not "smooth" when atomic dimensions are the basis of appraisal. It may be that the aluminum deposit of smaller weight presented a larger surface area of aluminum for oxidation. In any event, the only assumption that can properly be made is that the surface area of an aluminum film is close to, or equal to, that of the quartz plate on which it is condensed. Assuming a surface area factor of 1 and a density of 3.4, calculation shows the thickness of the oxide film which gave an increase of 5.2×10^{-6} g. per 16 sq.cm. is on the order of 10 Å.

Rate by Optical Methods

Other investigators have employed optical methods to measure the rate of oxidation of thin films of aluminum produced on glass or quartz by the same evaporation and condensation technique. (This seems to be the only method so far developed of securing aluminum initially free from oxide.) Optical measurements have the advantage of speed and are nondestructive, but calculation of thickness involves uncertain physical constants of the metal and oxide.

Hass[3] employed the Drude polarization technique for estimating the thickness of the oxide film formed on aluminum. He reports that an oxide film about 20 Å thick formed a very short time after exposure to normal air. Further growth was very slow and finally ceased after about six months, when it was about 90 Å thick. A second film on a smooth, mirror-like surface ceased to grow after about one month's exposure to normal air, at which time its thickness was about 45 Å.

Steinheil[4] and later Cabrera and Hamon[5] employed thin condensed films of aluminum and measured their opacity to light in order to determine the rate of oxidation. In this technique the conversion of aluminum to transparent oxide increased the transmission of light through the sample.

Steinheil reported a thickness of 400 Å for the oxide film after several months' growth. This is a considerably greater thickness than found by Hass. On using what he considered a more accurate value for the absorption coefficient for aluminum, Hass recalculated Steinheil's data and obtained a value of 77 Å for the oxide film, which was more in accord with his own observations.

Cabrera and Hamon, also employing the light transmission technique, determined the effect of humidity on the rate of oxidation. After five days' exposure to dry air (zero humidity), the thickness of the oxide film was about 15 Å; after 25 days it was 20 Å. With air at a humidity of 100%, the corresponding thicknesses were 25 and 50 Å., respectively. During the longer test period, the rate of oxidation appeared to increase slightly with an increase in the humidity of the air.

Oxidation at Elevated Temperatures

While the rate of film formation on aluminum at ordinary temperatures is of interest and importance from both theoretical and practical viewpoints, a similar importance attaches to the rates of formation at elevated temperatures. Aluminum products are customarily annealed in mill or fabrication shops at temperatures of the order of 350° C. (660° F.) for as long as 12 hr. Alloys may also be subjected to heat treatments for various periods at temperatures as high as 520° C.

[3] "The Growth and Structure of Thin Oxide Films", by G. Hass. *Optik*, V. 1, 1946, p. 134.

[4] "Structure and Growth of Thin Films on Metals After Oxidation in Air", by A. Steinheil. *Annalen der Physik*, V. 19, 1934, p. 465.

[5] "On the Oxidation of Aluminum in a Humid Atmosphere", by N. Cabrera and J. Hamon. *Comptes Rendus*, V. 225, 1947, p. 59.

(970 °F.). Since the rate of oxidation increases with increasing temperature, thicker and more protective films will form at the elevated temperatures employed for the annealing and heat treatment of aluminum.

Films that form during such treatments are much heavier than the film formed in air at ordinary temperatures and are approximately from 200 to 1000 Å thick. These thicknesses are estimated from the fact that some of them show interference colors, and that they can be stripped from the metal and examined under the electron microscope.[6] Figure 2 is an electron micrograph

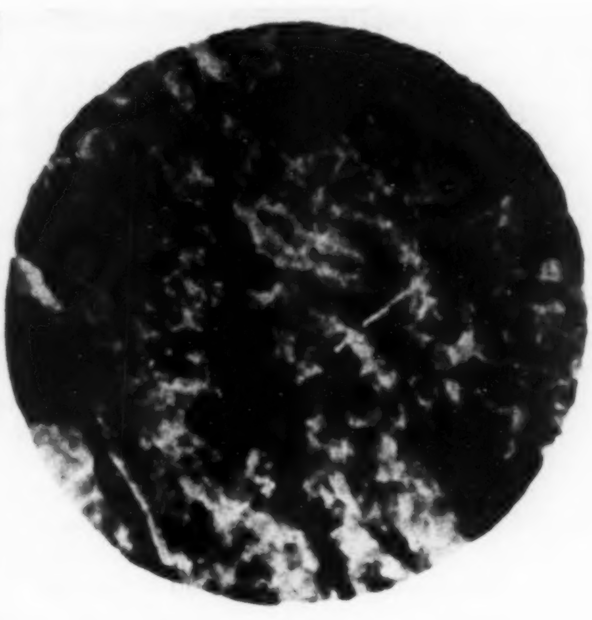


Fig. 2 — Electron Micrograph of the Oxide Film That Formed on a High-Purity Aluminum-Silicon Alloy Sheet During Heat Treatment for 24 Hr. at 570° C. (1050° F.). The oxide film was stripped from the sample and used for the electron microscope specimen. 20,000×

of the oxide film that formed on heat treating an aluminum alloy containing 0.9% silicon for 24 hr. at 570° C. (1050° F.). This film is rather porous and it cracked during the stripping process.

Two of the pioneers in measuring the rate of oxidation of aluminum at elevated temperatures were Pilling and Bedworth.[7] They advanced the idea that when the oxide film occupies a greater volume than the metal from which it is formed, the oxide film will tend to be compressed and consequently will be more compact in structure. Aluminum is one of the metals that forms a compact film of this type. Pilling and Bedworth found, on heating an aluminum wire in dry oxygen at 600° C. (1110° F.), that a film tended to form gradually on the surface. Growth apparently ceased after heating 60 to 80 hr.; no further increase was noted after 900 hr. These observations are of interest since they indicate quantitatively the formation of an oxide film of the barrier layer type which appears to be completely protective against continued oxidation. It should be noted, however, that the tests were made with dry oxygen. If air containing water vapor had been used, thicker oxide films would have formed.

A recent and comprehensive investigation of the weight gain of aluminum during oxidation at elevated temperatures has been made by Gulbransen and Wysong.[2] Using Gulbransen's microbalance, they measured the rate of oxidation of 99.98% purity aluminum at various temperatures up to 550° C. (1020° F.). Each test sample was a strip of 10-mil aluminum sheet with an area of about 11 sq.cm., weighed about 0.36 g., and had an air-formed film on its surface. Prior to starting a test, the sample was heated *in vacuo* at 475° C. (885° F.) for 30 min. in order to "degas" the specimen.

The results indicate that at 400° C. (750° F.) the rate of oxidation was slowing up decidedly after 2-hr. exposure to dry oxygen, while at 500° C. (930° F.) the rate of oxidation was still approximately linear with respect to time. Other tests at the latter temperature for 20 min. showed little change in rate with oxygen pressures ranging from 0.076 cm. to 7.6 cm. of mercury — a hundredfold change in oxygen concentration.

One of the most striking results of these tests was the wide variation in the effect of surface condition on the rate of oxidation. For example, one sample, which had been abrasively polished, showed a weight gain of 22×10^{-6} g. after heating for 100 min. at 500° C. (930° F.). Another sample from the same roll of aluminum foil showed a weight gain of only 10×10^{-6} g. under the same conditions. The samples which had been abrasively polished were much more active toward oxygen than those which were merely cleaned by washing with petroleum ether and alcohol. The rate of oxidation on the first of these samples after 100-min. heating was 5 times as rapid at 500 as at 400° C. All of these samples, of course, had an air-formed film on them at the start of the test. Efforts to produce a uniform, low-oxide surface by abrasive polishing, by electrolytic brightening,

[6] "Application of Electron Microscope to Study of Aluminum Alloys", by F. Keller and A. H. Geisler. *Transactions* of the American Institute of Mining and Metallurgical Engineers, Institute of Metals Division, V. 156, 1944, p. 82.

[7] "The Oxidation of Metals at High Temperatures", by N. B. Pilling and R. E. Bedworth. *Journal* of the Institute of Metals, V. 29, 1923, p. 573.

and by chemical treatment produced significant changes in the activity of the surface, but the interpretation of the results in terms of surface characteristics is very difficult.

Rate of Oxidation as Determined by Peroxide Formation

J. R. Churchill of Alcoa's Aluminum Research Laboratories[8] has suggested that the "Russell effect", a reaction involving the formation of hydrogen peroxide by the action of moist air on certain metals, might be employed to measure the rate of oxidation of aluminum. The original experiments indicating that hydrogen peroxide was formed during the oxidation or corrosion of metals were made long ago by Schönbein in 1858. Colson, however, was the first to observe that certain metals — zinc, cadmium and magnesium — could affect silver haloid emulsion in the dark, which he attributed to the formation of a metallic vapor. Thorough investigations by Russell between 1897 and 1908, and published in the *Proceedings* of the Royal Society, led him to the conclusion that the photographic images Colson obtained were caused by emanation of hydrogen peroxide from these metals, but other investigators did not agree with these conclusions.

In 1939 Churchill published the results of his new experiments on the Russell effect. He demonstrated that when a stream of moist air, free of peroxide, was passed through a column of freshly cut aluminum turnings, H_2O_2 was formed in appreciable quantities, as shown by the yellow color developed when the air was subsequently bubbled through a solution of titanium sulphate. He also determined qualitatively that hydrogen peroxide formed most rapidly with freshly prepared surfaces of aluminum, and less rapidly with surfaces containing various thicknesses of an air-formed oxide film. His experiments indicated that the peroxide was formed by the oxidation of atomic hydrogen by molecular oxygen. In his experiments, test pieces of aluminum were placed against a sensitive photographic emulsion and after various exposure periods the latent image was developed. Tests of this type showed that the intensity of the photographic image decreased appreciably as a protective oxide film formed on the aluminum sample.

This procedure for measuring the rate of oxidation was found in our laboratories to be more complicated than first anticipated. For example, the amount of hydrogen peroxide liberated from a single test piece was small and insufficient to cause adequate darkening of the photographic emulsion in a short exposure. A technique was developed, however, by which a number of test pieces could be exposed for the desired periods, and this yielded interesting results. Churchill also discovered that a fourfold denser image could be obtained by aging the photographic plates for 40 to 60 hr. before developing.

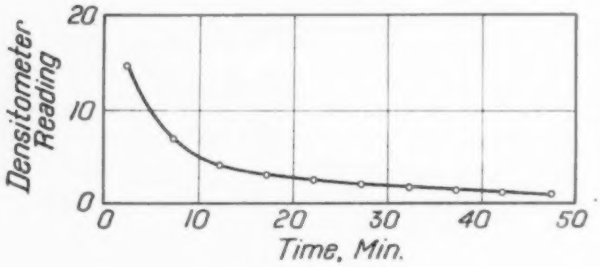


Fig. 3 — Densitometer Readings on Photographic Images Produced by Emanation of Hydrogen Peroxide From High-Purity Aluminum Sheet at Various Stages During Forming of a Natural Oxide Film in Moist Air. Each point represents the effect from 10 specimens and is for the elapsed time after removing old oxide film, plus half the exposure time for each sample. (J. R. Churchill, Aluminum Research Laboratories)

Figure 3 shows the results of measurements made on aluminum sheet of high purity (99.95%) by the following procedure: A series of 10 test specimens was prepared and a slot-like holder arranged above the photographic plate. The first specimen was freshly surfaced by abrasion and quickly placed in position No. 1 in the slot. After the exposure of 5 min., the second specimen was abraded in the same way, inserted in the No. 1 position, and the first specimen pushed over to the second position. In this way each specimen, after a definite period of aging and exposure to the plate, was pushed to a new position on the plate, so that successive images were obtained, representing the cumulative exposure of 10 test pieces which had aged a definite number of minutes from the time of surfacing. After the series was complete the specimen holder was removed, and the plate was allowed to age for 24 hr. and then was developed. The density of the image was measured with a densitometer such as used for spectrographic work. In the graph, Fig. 3, the galvanome-

[8]"The Formation of Hydrogen Peroxide During Corrosion Reactions", by J. R. Churchill; *Transactions* of the Electrochemical Society, V. 76, 1939, p. 341. R. Colson, *Comptes Rendus*, V. 122, 1896, p. 598; V. 123, 1896, p. 49. W. J. Russell, *Proceedings* of the Royal Society, V. 61, 1897, p. 424; V. 63, 1898, p. 102; V. 64, 1899, p. 409; V. 78, 1906, p. 385; V. 80, 1908, p. 376.

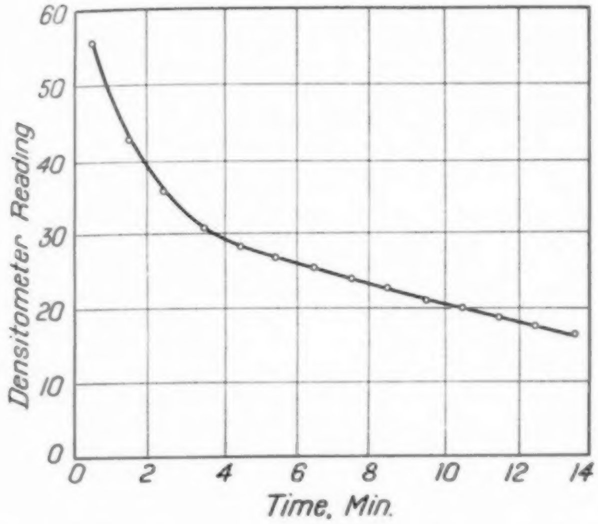


Fig. 4 — Densitometer Readings on Photographic Images Produced by Emanation of Hydrogen Peroxide From Surface of Commercially Pure Aluminum Sheet at Various Stages During Forming of a Natural Oxide Film in Moist Air. Each point represents the effect from 30 samples and is for the elapsed time after removing the old oxide film, plus half of the exposure time (0.5 min.) for each sample

ter deflection on the densitometer was taken as a measure of the density of image and plotted as a function of the time the aluminum test piece had oxidized. (The point plotted on the time axis is the elapsed time since surfacing plus half of the exposure period. For example, the first sample was placed on the plate just 5 min. after surfacing; this period plus half of the exposure period places the densitometer observation at 7½ min. on the graph.)

In Fig. 4 is shown the photographic effect of peroxide evolution from commercial 2S sheet (99.2% Al). In this graph the observation periods are spaced 1 min. apart and each point on the graph represents the effect of exposure to 30 test pieces. The cumulative effect of the exposure of the photographic plate to each of the 30 test samples gave densitometer readings which form a very uniform curve.

No definite proportionality has been established between the depth of oxidation of the aluminum and the amount of hydrogen peroxide produced (and its effect on the density of image on the photographic plate after development). Nevertheless, these two curves present an interesting picture of this reaction in the early stages of formation of an oxide film on aluminum. They show it is only a matter of minutes until the rate of reaction becomes quite small. (Other tests have shown that samples of commercial aluminum sheet give an identifiable image by this technique when tested 43 days after abrasion. Another test piece gave a distinct but considerably weaker image when tested 79 days after surfacing!)

The formation of hydrogen peroxide apparently results from the oxidation of active hydrogen, such as hydrogen ion or atomic hydrogen. This hydrogen is formed by the reaction between aluminum and water while the oxide film is building up on the aluminum. There seems little chance that water molecules can diffuse through the oxide lattice so that, except for substantial defects in the continuity of the film, little water vapor can penetrate to the metal once a barrier layer of oxide is formed. This makes it appear probable that aluminum ions are diffusing to or near the surface where they react with adsorbed water vapor.

In interpreting these data in terms of film formation, the possibility should also be kept in mind that certain alloying elements, such as copper and manganese, may act as catalysts and be effective in decomposing hydrogen peroxide. Such action would, in effect, reduce the "Russell effect". However, experiments with alloy 17S (4% copper, 0.5% magnesium, 0.5% manganese) give curves quite similar to those shown for pure aluminum and do not indicate any important effect of alloying constituent.

Summary

The natural oxide film on aluminum has long been known to be responsible for the resistance of aluminum to weathering and oxidation in general. This film starts to form at once when an aluminum surface is exposed and reaches a protective thickness in a matter of minutes. In dry oxygen, the film on bare aluminum grows rapidly at first and stops abruptly after a period of some 10 to 40 min. At this time the film has a thickness corresponding to a few atom layers. In the presence of water vapor, growth will continue slowly for a substantially longer period. The natural oxide film, in air, may be said to have a thickness ranging from 10 to 100 Å. At higher temperatures, particularly above about 400° C. (750° F.), growth is much more rapid, and the ultimate thickness of film considerably greater. In the presence of water vapor, growth of the film is attended by the liberation of a small amount of hydrogen peroxide and the effect of this hydrogen peroxide on a photographic emulsion may be employed to follow the reaction.

(Composition and properties of the natural oxide film on aluminum will be discussed in a later paper.)

Tests at Eniwetok on Improved Atomic Bombs

ON MAY 18 a statement was issued at Honolulu, by the Secretary of Defense, containing interviews with the army, navy and civilian chiefs of "Joint Task Force Seven", which had constructed a proving ground at Eniwetok Atoll in the Pacific and had recently completed the first series of tests of atomic weapons known as "Operation Sandstone". Verbatim excerpts follow:

Construction of the proving ground commenced in late December 1947. In general this involved construction and rehabilitation of living facilities for the scientific group and troops, installation of utilities, storehouses and shops. It was necessary to repair air strips to permit landing of heavy planes, including C-54's, B-29's and B-17's, as well as dockage for small water craft. Construction of the actual testing ground involved placement of various scientific and test equipment, and very extensive radio and telephone services, including nearly one million feet of submarine cable. Some 50,000 tons of material were shipped to the test area. The over-all strength of the Joint Task Force was approximately 9800, including civilians.

All of the wartime bombs, including those at Bikini, were about on a par as far as their state of development was concerned. They were the wartime weapon—designed under extreme pressure and without regard for many problems which in the long run are of great importance in the military application of atomic energy.

These tests at Eniwetok during April and May were to determine how the bombs under development by the Los Alamos Scientific Laboratory of the Atomic Energy Commission during the past two years would work, and to determine their efficiency. The test program involved a series of nuclear explosions, carried out under conditions as close to laboratory control as we could make them, and with very extensive instrumentation.

The work fell into two categories. The one was to fill in gaps in the knowledge gained at Bikini. The purpose of the Bikini tests was to test the effect of nuclear explosion on naval equipment and other materials, as well as on animal and marine life. Many of the tests made at Eniwetok to fill in the gaps of our knowledge in these respects gained at the Bikini tests (and made by several agencies of the armed forces) were not very extensive. However, the desired data were obtained. We did not conduct the postponed Bikini deep-underwater test.

The second category of tests and experiments was designed to answer questions arising in connection with the military applications of atomic energy. The ultimate purpose of the program is to insure efficient utilization of the national resources required for the development and application of atomic energy. It is obvious that a research and development program of new weapon designs cannot long be fruitful if the product of the program never gets tested. If the United States elects to develop and manufacture atomic weapons, these weapons must be tested. Unlike other bombs, however, the cost in actual cash, man-hours, and natural resources is quite high for each weapon. Morever, the physical processes going on during the explosion of an atomic bomb are very complicated. For these reasons, development and improvement of atomic weapons cannot be carried on by the common methods of making small changes in current models and proof-testing after each change.

Proof-tests of new models often can be carried out under conditions that make it possible to attain secondary, but important objectives. Without interference with the primary objective, much information can be gained which is useful in the peaceful applications of atomic energy.

A very great deal of physical research and mathematical analysis goes into the plans of an atomic weapon. Therefore, tests of the kind we have just completed are designed primarily to provide experimental data necessary for a better understanding of the process of nuclear explosion, and necessary to form a sound basis for improved design of weapons. Certainly such tests do include proof-firing new models of weapons, but the model types must be selected carefully in order to make information obtained from one test supplementary to that obtained from another. A well-planned series of atomic weapons tests can yield much more information than an equal number of unrelated single tests.

Our tests were not successful merely because the weapons we used exploded with a loud bang. They were successful because the weapons did explode and we obtained good experimental data which will guide us in research and development in the future. One of the most gratifying results of the entire operation has been the confirmation of the large body of ideas, theories and methods which have grown out of the theoretical and experimental work done since the war at Los Alamos.

[In view of the friction between civilians and soldiers that often appeared in America during the wartime activities, the following by Dr. D. K. Froman, the scientific director of the task force, may be significant:] Throughout the whole life of the Joint Task Force there has not been a single incident which impeded any test or measurement and which arose from the rather great differences between military and civilian philosophies and methods of operation. A year ago I would not have believed such a working relationship could be achieved.

Presentation of verbatim extracts from important contemporary documents concerning atomic energy does not imply that the Editor agrees with the opinions quoted, nor that they are expressions of A.S.M. policy.

Impact Testing of Weldments

By William C. Long
Experimental Metallurgist
Delco Products Division
General Motors Corp.
Dayton, Ohio

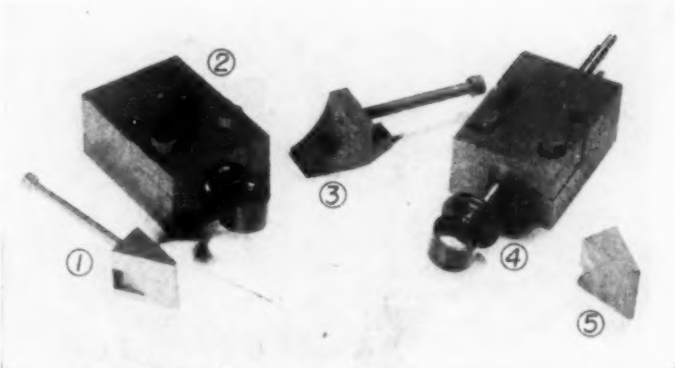


Fig. 1 — Striker Noses and Holding Devices for Testing Welded Assemblies in Riehle Impact Machine

DURING the reconversion period of late 1945 and early 1946, after the last war, the metallurgical laboratory of the Delco Products Division, General Motors Corp., was requested to develop a test whereby the weld quality of small steel assemblies, electrically butt welded, could be determined. At that time a tensile test was used to measure the strength of assemblies welded in production, and it was noted from time to time that although the tensile requirements were met, certain welds were brittle, porous, and lacked the proper degree of fusion.

The assemblies to be tested were small parts of the Delco automobile shock absorber (direct-acting type), and consisted of two basic components, namely, *(a)* a piston rod made of S.A.E. 1040 steel, threaded on one end and the other end press fitted into a round cover plate, and *(b)* an oil reservoir cup, more or less dished depending on the design. The latter is stamped out of cold rolled strip; its function is to form the base of the shock absorber cylinder. A pair of these components form opposite ends of the shock absorber, and in order that they may be attached to the automobile it is necessary to weld — either to the cylinder base cup, or to the piston rod — one of the following devices:

1. A hexagonal-headed stud (S.A.E. 1010 steel).
2. A ring made of hot rolled strip steel.
3. A mild steel bayonet (cold headed).

View 4 of Fig. 1 shows one such assembly in a testing fixture. The piston rod with its dished cover plate extends through the fixture; to its near end is welded the ring formed of hot rolled steel strip.

During operation of the shock absorber in the automobile, these small welded assemblies are subjected to bending and impact stresses, as well as tension, and in view of these stresses an impact test was considered feasible. With this thought in mind, special holding and striking devices were designed for use in a standard Riehle impact testing machine, and they were used in the preliminary investigation. These are shown in Fig. 1. Item 1 shows the first-used striker and the bolt for its attachment into the swinging tup of the Riehle machine; Item 5 shows a later design, whereas Item 3 is a striker with offset nose for testing assemblies to which are welded either studs or bayonets. The noses of all these strikers are flat.

Item 2 of Fig. 1 shows an assembly in a holding device which can be bolted to the anvil of the Riehle machine, whereas Item 4 is the same (rod, cover plate and ring) with the assembly moved out of its proper position to show the boss on the fixture for supporting the piece as close to the plane of the weld as possible.

All of the holders and strikers were fabricated from a silicon-manganese steel, hardened, drawn and multiple tempered to a hardness of C-52 to 55.

All four of the attaching bolts for the fixtures required special threads, and they were hardened and drawn to maximum impact strength.

As testing progressed, the following facts were revealed:

1. No apparent correlation exists between impact and tensile test results on weldments. Several hundred companion tests were conducted on identical specimens welded during a 48-hr. production run on several resistance type welding machines, but no relationship was found to exist between the tests. Sound welds of proper area had good tensile tests and impact strength; welds of obviously inferior quality (as judged by area or condition of fracture) had acceptable tensile strength but were quite deficient in toughness.

2. The impact energy absorbed, measured in ft-lb., and the weld fracture appearance were found to be reliable indicators of general weld quality, fusion, porosity, ductility and alignment in the electrodes.

Valid and reproducible impact test-results were made possible by reason of the fact that the weld areas of the parts were fabricated and welded in mechanically uniform processes. Measurements on tested specimens revealed that the areas of fusion could be maintained within ±0.020 sq.in. The results showed that the impact test was a reliable yardstick to measure quality of production weldments.

When laboratory experiments had progressed sufficiently, several sturdy impact testers of the "guillotine" type were designed and built, and are now being used by the inspection department on the production lines (Fig. 2). These machines consist of known weights to which are attached impact-resistant steel strikers; they fall through guide tubes from known heights, and impinge upon the rigidly supported assemblies to be tested. Weights and distances are based upon standards established from laboratory results. Testing in the production lines provides an immediate and constant check on weld quality, and is an invaluable aid during the periods when welding machines are being adjusted or new fixtures set up. It is interesting to note that the quality of production weldments in shock absorbers increased 200 to 400% during the first three months of impact testing, and has been maintained consistently at a high level during the past two years.

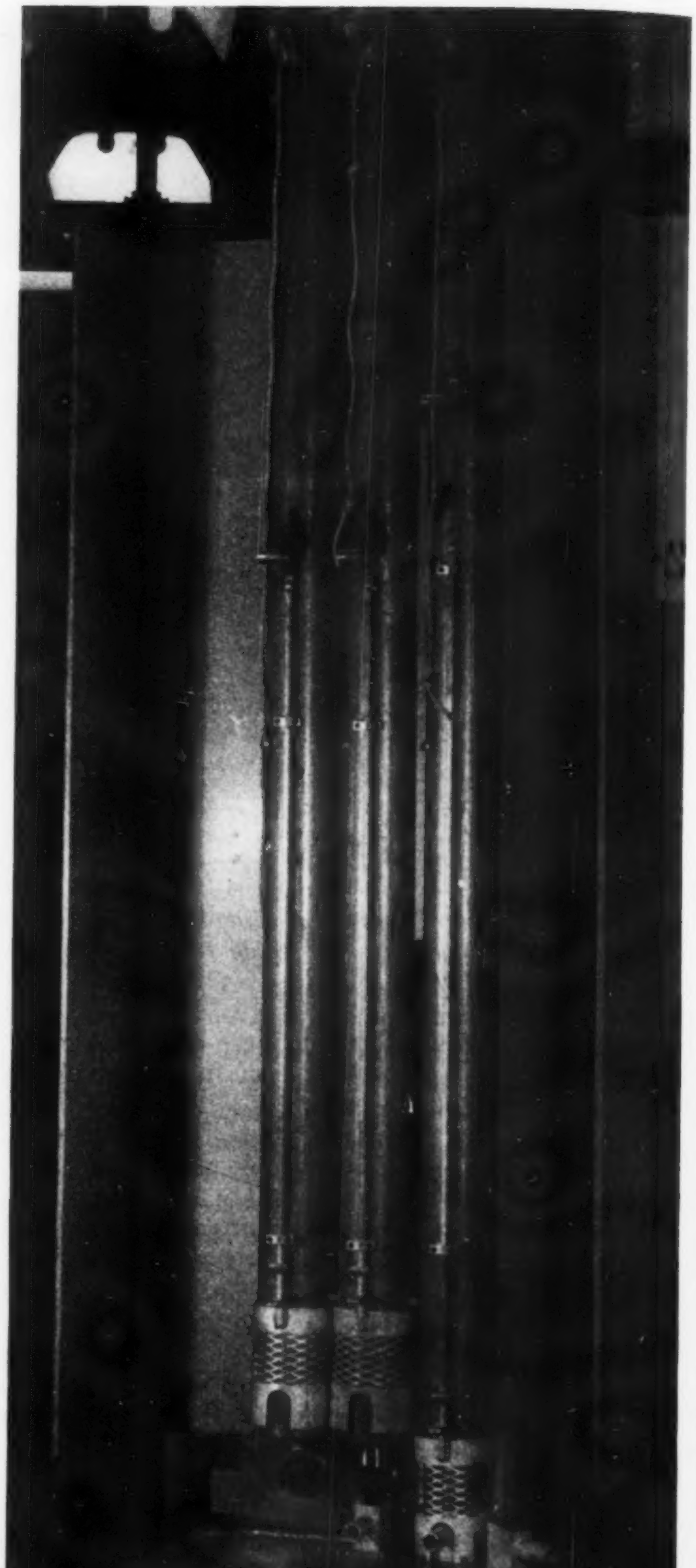


Fig. 2 — Triple Impact Machine With Fixtures to Test Three Different Assemblies in the Production Line

Figure 2 shows one such machine in a sheet steel booth. It is in fact three individual drop-impact machines, identical except for tups and fixtures. The latter are adapted for two sizes of assemblies, left and center (piston rod, round cover plate and ring), and the right-hand one is for an assembly of rod, plate and bayonet. The photograph also shows various test sequences.

The striker weight is raised within the guide tube by the lift rope. The height for the required striking energy is fixed very simply by an internal ratchet attached to the striker release lever on the front of the tube (about one third the way down from the top). The perforated safety guard at the bottom of the tube is also raised; it is attached to a rod easily sliding through cleats on the front of the guide tube, and a latch at top interlocks into the striker release so the weight cannot fall unless the safety guard is in proper "down" position, as shown in the right-hand unit.

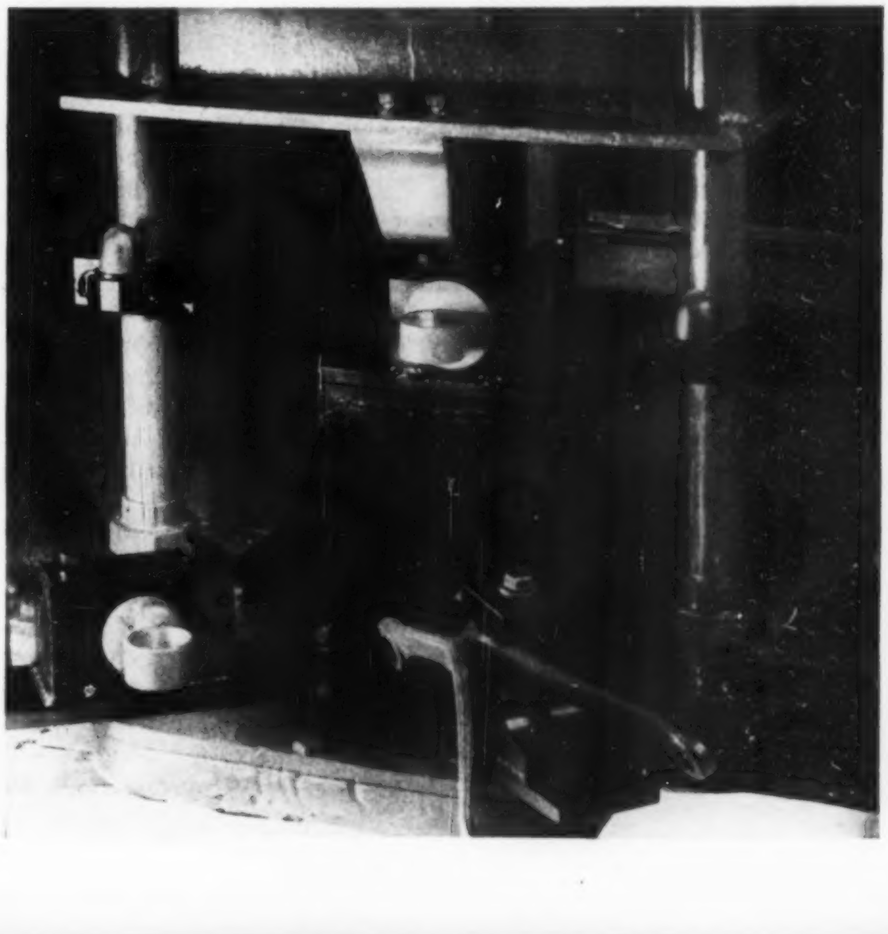
The piston rod of the assembly to be tested is then slipped into a collet in the base plate and a bottom stop or anvil placed in position just clearing the underside of the assembly (center unit). Next, the perforated safety guard is dropped into "down" position (right-hand unit) by depressing the small lever just under the lower cleat. Its support rod also drops, clearing the striker release. A light touch on this release frees the weight and the striker assembly drops down, hitting the assembly to be tested. (In the center unit the safety guard is up, the stopping anvil is in position, and the striker is shown as if in motion at the very instant of impact on the test piece. The unit at right shows the actual position of the safety guard at that time.)

In routine inspection, a representative sample is tested by several blows, whose energy is fixed by the height of fall of the striker assembly. The first blow should not break the assembly. Successive drops break the weld, and it is then examined visually for area, uniformity and character of fracture. Striking energy for tests on all varieties of shock absorber parts are determined by the laboratory, having regard for the size of the weldment and its specifications.

Recently it was found necessary to construct a large capacity guillotine-type machine (Fig. 3) because of increased size and strength of parts and welds. One of the more interesting weldments now being investigated with its aid consists of two S.A.E. 1035 forgings which are butt welded together at two side lugs, thus forming the shock absorber arms. This welded unit is shown in its holding fixture at lower center of Fig. 3, tilted somewhat to show the weld clearly. Tests on this type of weld (and also on the smaller units involving the welds whereby studs or bayonets are attached) require much more impact energy for fracture because they are tested under a pure shear condition. Some mechanical advantage rather than pure shear is desired in the test to break the welds properly, but shear without bending is also necessitated by the shape of some of the parts that are to be tested.

In Fig. 3 a rod-disk-ring assembly is in a fixture on the base plate at left, and another is in the test position on top of the anvil in the center of the view. Immediately above is the nose of the striker, bolted to a crosshead. The top of the picture cuts off most of this crosshead, which has provision for attaching weights as desired. It is raised by a small electric hoist, and slides up and down on side rods — really pipes which extend from floor to ceiling. Striker plates with rubber bumpers, which are bolted to the guide rods, serve to stop the tup immediately after the test piece is broken.

Fig. 3 — Guillotine Type of Drop Testing Machine for Heavier Weldments Beyond Capacity of Fig. 2



Proper Frequency for Induction Heating of Nonmagnetic Metals

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Cleveland

All materials (magnetic or nonmagnetic, conductive or nonconductive) can be heated by electromagnetic induction in appropriate electrical apparatus. At the present time, commercial power sources are available for heating, by this phenomenon, all materials which are normally considered as electrical conductors, but the following paper concerns the heating of nonmagnetic materials such as copper, aluminum or austenitic steels. However, it is to be remembered that steel, a magnetic material, loses its magnetic properties in the temperature range used for hardening or forging, so the topic has more scope than might be apparent from the title.

All induction heating equipment consists of a source of high-frequency power, metering devices, contactors, automatic timing controls, suitable transformers and capacitors, and inductors, plus quenching auxiliaries, if needed for hardening. The inductor may vary from a single turn of copper wire to several turns of copper tubing. Design details are generally a matter of specific application and the users of induction heating equipment should take advantage of services and recommendations of equipment manufacturers.

There are four sources of high-frequency current which find commercial acceptance for induction heating:

1. Rotating equipment (motor-generators) of frequencies from 1000 to 30,000 cycles are used at capacities up to 10,000-kw. rated output. Most commonly used frequencies are 960, 3000, 10,000 and, quite recently, 30,000 cycles. Equipment sizes start with small packaged units at 7½-kw. rated output.

2. The spark gap oscillator provides frequencies of the order of 30,000 to 400,000 cycles, and rated inputs up to approximately 40 kva. Normal operating efficiency gives maximum output of approximately 25 kw. for the largest unit commercially available.

3. Frequencies of several hundred thousand cycles (usually around 500,000) are developed by vacuum tube oscillators. Power output is again a limiting feature with 10, 20, 25, and 50 kw. appearing in standard equipment now available. Frequencies of as high as 3,000,000 cycles are also used for special applications, but power output is also limited. Units at greater power ratings can be procured, but at high cost per kilowatt and large space requirements.

4. A recently developed power source is the mercury arc converter used to convert 60-cycle input to an output of from 500 to 1500 cycles. Conversion is accomplished at high efficiency. At present, the output frequency is limited to about 1500 cycles.

With any of these power sources, for a given inductor and load, the amount of electrical energy producing heat in the load relative to the total energy input to the inductor is dependent on the frequency of the latter. In other words, optimum energy conversion can be obtained by proper selection of frequency. It will be shown in this paper that, when heating solid cylindrical loads, a range

will be reached as the frequency is increased where higher frequency will result in negligible change in energy conversion. It will also be shown that there is a particular frequency at which maximum energy conversion is obtained when heating hollow cylinders, and any increase or decrease in frequency from this value will result in less efficient conversion.

In addition to efficiency, there are other important factors to be considered which influence the selection of frequency. These are metallurgical (heat pattern) and economic (cost of equipment).

Heating Efficiency

In order to investigate the effects on efficiency of current frequency and the material's size, shape and resistivity, it is necessary to consider the characteristics of the fundamental circuit. An inductor and load combination is shown in Fig. 1.

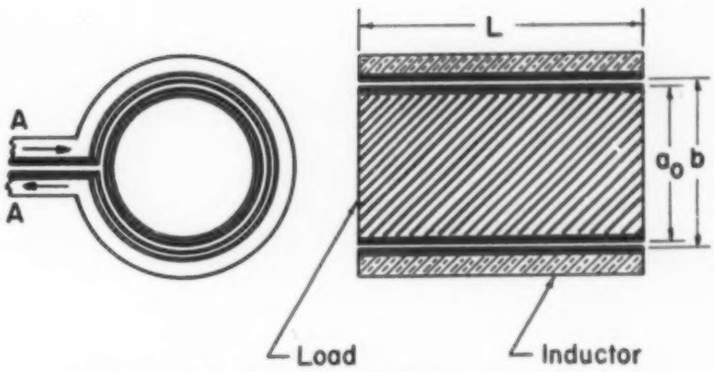


Fig. 1 — Sketch of Ideal Inductor and Load

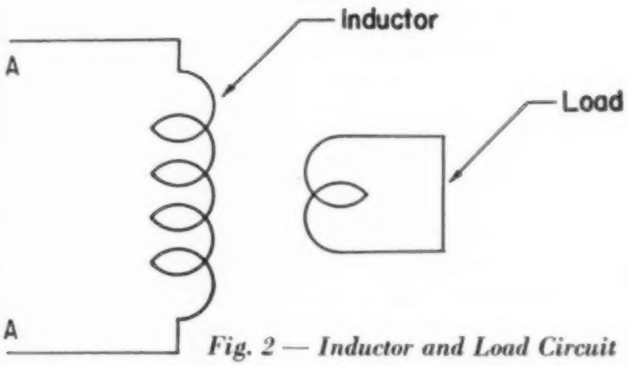


Fig. 2 — Inductor and Load Circuit

Looking at the end view, the depth over which the electrical energy is concentrated is shown by the shaded portions. This assumes a sufficiently high frequency to produce these relatively shallow "energy depths". The following equations give the depth in inches within which 90% of the electrical energy is concentrated, provided this depth is small compared to the load diameter a_o.

$$\text{For inductor coil, } d_p = 3160 \sqrt{\frac{\rho_p}{f}} \qquad (1)$$

$$\text{For load, } d_s = 3160 \sqrt{\frac{\rho_s}{f}} \qquad (2)$$

Where ρ_p = the resistivity of the inductor in ohm-inches
ρ_s = the resistivity of the load material in ohm-inches
f = frequency of the current in the inductor in cycles per sec.

The circuit for the inductor and load combination is shown in Fig. 2, and an equivalent electrical circuit for Fig. 2 is shown in Fig. 3, wherein

R_p = resistance of the inductor coil
X_p = reactance due to the flux between inside and outside radius of inductor coil
X_o = reactance due to the flux in air gap between inductor and load
X_s = reactance due to flux within the load as reflected in the inductor coil
R_s = resistance of load, as reflected in the inductor coil

If current is passed through the circuit of Fig. 3, electrical power will be dissipated in the form of heat in the resistance elements only, that is in R_p and R_s. Fundamentally, electrical power P in watts is given by the well-known equation $P = I^2R$, where I is the current in amperes and R is the resistance in ohms.

Applying this equation to the circuit of Fig. 3, the power dissipated is as follows:

Total power input, $P_T = I^2(R_s + R_p)$
Load power, $P_s = I^2R_s$
Inductor power loss, $P_p = I^2R_p$

Concerning ourselves with the efficiency η of power conversion, this is defined as ratio of the power dissipated in the load to the total input to the inductor. Therefore:

$$\eta = \frac{I^2R_s}{I^2R_p + I^2R_s}$$

$$\text{or } \eta = \frac{R_s}{R_p + R_s} = \frac{1}{1 + R_p/R_s}$$

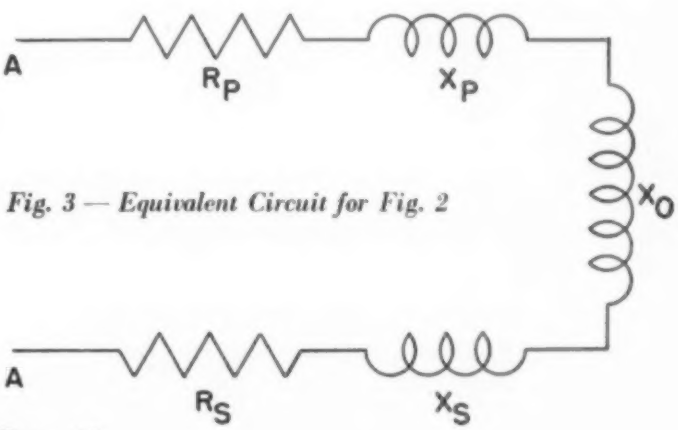


Fig. 3 — Equivalent Circuit for Fig. 2

Therefore, the efficiency depends inversely on the ratio R_p/R_s, increasing as the ratio decreases or as R_s (the resistance of the load) becomes large compared to R_p (the resistance of the inductor coil). Consequently, to determine the electrical efficiency, we need only know this ratio of R_p to R_s.

For the inductor and load shown in Fig. 1, with axial length L long compared to inductor diameter b, we have

$$\frac{R_p}{R_s} = \sqrt{\frac{\rho_p}{\rho_s}} \times \frac{b}{a_o} \times \frac{1 + d_p/b}{K_2} \qquad (3)$$

where K_2 = resistance coefficient for load according to Fig. 4, and a_o is the diameter of the load (Fig. 1). This equation relating the resistance of load and inductor coil applies to solid cylindrical loads and to hollow loads for a wall thickness equal to or greater than twice the depth d_s.

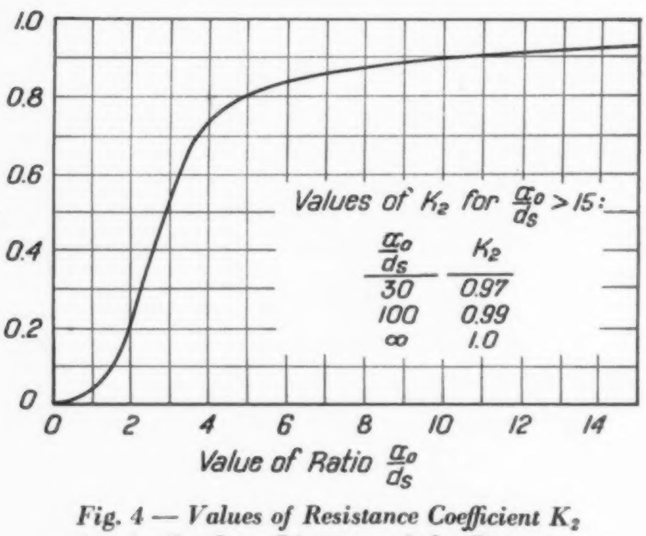


Fig. 4 — Values of Resistance Coefficient K_2 for the Load as Diameter of the Piece Becomes Large in Proportion to Heated Depth

Observing this last equation and realizing that the percentage of inductor input that is dissipated in the load is dependent on the ratio R_p/R_s, it is seen that for an inductor and load of given dimensions and material resistivity, the only factor that varies with frequency is $\frac{1 + d_p/b}{K_2}$. Further, it is found that K_2 depends on a_o/d_s, the ratio between diameter of the piece being treated and the treated depth, and approaches 1 as this ratio approaches infinity. For a given diameter of load, a_o, the ratio a_o/d_s approaches infinity as frequency is increased, since d_s is inversely proportional to the square root of the frequency, as noted in equation (1). On the other hand, it is noted from equation (1) that d_p, the energy depth at the inductor coil, *decreases* as frequency is increased. Therefore, at a sufficiently high frequency, d_p/b becomes negligibly small, K_2 is approximately 1, and the third factor of equation (3) approaches unity. In this range further increase in frequency will result in negligible gain in efficiency.

This fact is shown in Fig. 5, where electrical efficiency is plotted against frequency when heating aluminum rods of ¾ and 2 in. diameter, respectively. It is noted that after reaching a frequency of approximately 4000 cycles the efficiency of heating the smaller rod remains relatively constant with further increase in frequency. Obviously, for a commercial installation, the frequency used would be 3000 cycles, since that is obtainable from readily available induction heating equipment. The efficiency of power conversion (point A) is 0.59. It would reach only 0.64 at 500,000 cycles! As the diameter is increased for a given material, or a certain resistivity, the frequency at which efficiency begins to level off is reduced. This is shown by comparison of the two curves in Fig. 5; the upper curve for the 2-in. aluminum rod is practically flat throughout.

As the resistivity of the material increases, the frequency at which efficiency begins to level off increases. This is shown in Fig. 6, which gives curves for a load of relatively low resistivity (aluminum at 1100° F.) and one of high (Type 302 stainless steel at 2200° F.). Aluminum represents one of the lowest resistivity metals (4.5×10^{-6} ohm-in.) and stainless steel one of the highest (54×10^{-6}). Other common metals such as magnesium, brass, zinc, tin, and lead fall in between.

For hollow rods or tubes there is generally a frequency for maximum conversion efficiency. This effect is shown by the curve in Fig. 7. The efficiency calculation for this curve is based on equations in an earlier paper by J. T. Vaughan and J. W. Williamson ("Design of Induction Heating Coils for Nonmagnetic Loads", *Electrical Engineering*, V. 64, 1945, p. 587). It is noted that

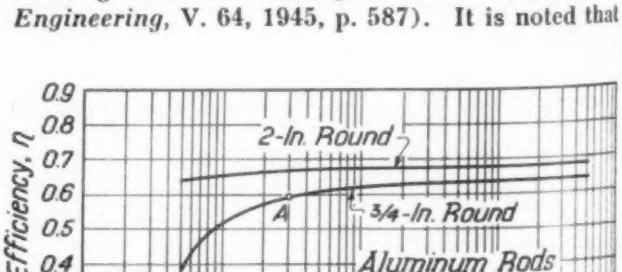


Fig. 5 — Relationship Between Efficiency of Power Conversion and Frequency of Alternating Current When Heating Aluminum in Inductors ¼ In. Larger Than Work

maximum efficiency of power conversion when heating the brass tube assumed in the computations occurs somewhere around 3500 cycles. However, the peak is quite flat and at the commercially available frequency of 3000 cycles the efficiency is only slightly less than the maximum. In the event that power requirements are below approximately 100 kw., a frequency of 10,000 cycles would be used at little loss of efficiency. A frequency of 500,000 cycles would involve a somewhat greater loss in efficiency (63% at *H* in place of 74% at *M*, the maximum). These frequencies are available from standard commercial units.

The frequency at which maximum efficiency for heating tubing will be obtained is

$$f = \frac{34.6 \times 10^6 \times \rho_s}{(a_o - t) \times t} \qquad (4)$$

where t is the wall thickness in inches and other

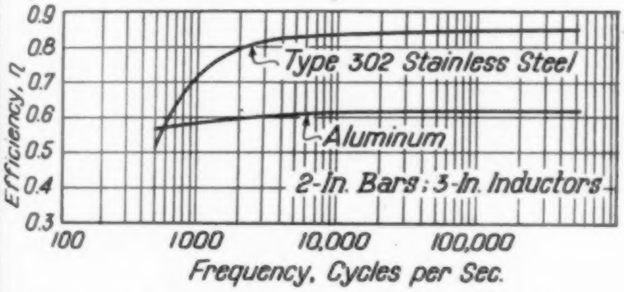


*Fig. 6 — As Resistivity of Metal Increases, Optimum Power Conversion Occurs at Higher Frequencies**

terms are as previously defined. This equation applies only to tubes or hollow rods with wall thicknesses less than 13% of the outside diameter.

Inductor Input

A further consideration in the selection of frequency is the inductor input in kilovolt-amperes relative to kilowatts. This importance is due to the fact that as the ratio of kva. to kw. increases, it is necessary either to increase inductor current or inductor voltage for a given performance (kilowatts). The ratio of kva. to kw. increases with frequency for a given inductor and load. As the inductor *current* is increased, difficulties are encountered in the transmission line between inductor and shunting capacitors. The inductor *voltage* should not be increased beyond a point

*In Fig. 5 to 8 the resistivity of the materials is taken as follows:

Copper inductors: 0.75×10^{-6} ohm-in.
Aluminum at 1100° F.: 4.5×10^{-6}
Stainless steel at 2200° F.: 54×10^{-6}
Brass at 1600° F.: 5.8×10^{-6}

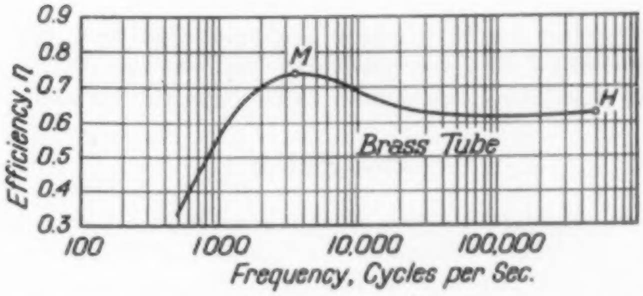


Fig. 7 — Tubing Is Most Efficiently Heated at Definite Frequencies. In the above, 1-in. brass tube is assumed, 0.06-in. wall. Copper inductor, 1.5 in. diameter

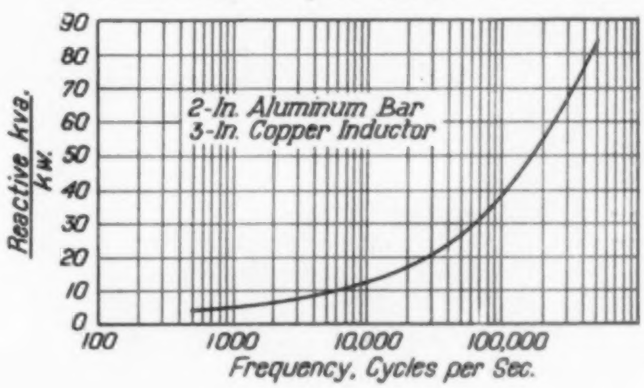
where it becomes hazardous to operators, necessitating undue precautionary measures and giving rise to insulation problems.

The reason for this can be seen by studying the equivalent circuit in Fig. 3 and its components. The ratio of reactive kva. to kw. is given by the proportion $\frac{X_p + X_s + X_o}{R_p + R_s}$

Usually X_o, the reactance due to the flux in air gap between inductor and load, is the largest reactive component. It varies directly with frequency. However, R_s and R_p (the resistances of load and inductor coil, respectively) vary approximately as the square root of frequency. Therefore, the ratio of kva. to kw. increases with frequency. This is shown graphically in Fig. 8.

Another effect that this ratio has on the selection of frequency is that the necessary capacitor in parallel with the inductor coil circuit for power factor correction (or tuning) is dependent upon the reactive kva. in the inductor coil circuit. On capacitors used for motor-generator set frequencies. the cost per kva. for 3000 and 10,000-cycle capacitors is about the same, while the cost for 1000-cycle capacitors is approximately double that for 3000-cycle. With further decrease in frequency, the capacitor cost rises rapidly. Thus

Fig. 8 — Relation Between Frequency of Heating Current and the Ratio of Its Reactive Kva. to Kw.



there is an economic balance between the increased cost for capacitors and decreased cost in generating equipment, as frequency is reduced.

Heat Pattern

The heat pattern desired is another important factor which influences the selection of frequency. If a solid bar is to be heated for forging, it would generally be desired to have a uniform temperature throughout the cross section. As the frequency is lowered, the depth of zone in which the electrical energy is concentrated increases, thus relying less on thermal conduction to equalize temperatures throughout and allowing a faster heating time for a given temperature uniformity.

The diameter of stock to be processed must also be considered, as it affects the ratio of surface area to volume and its response to minimum frequencies. Obviously, through-heating of sections several inches in diameter would not be economical in practice if very high frequencies were used, since the necessarily low energy input and resultant long heating time for the heat to flow inward by conduction (from surface layers heated inductively yet without overheating) would result in high heat losses to the surroundings by radiation and convection.

In commercial production there are numerous operations where one tube is brazed inside another, or a collar brazed on a piece of tubing. If too high a frequency is selected and an external inductor is used, all of the electrical energy will be confined to the outer tube; the heat reaching the inner tube will get there merely by thermal conduction, and the amount would vary with tightness of fit. In such instances, it would be more desirable to use a lower frequency which would allow a portion of the electrical energy to be dissipated in the inner tube as well, and therefore secure a greater degree of uniformity in the heating pattern.

This discussion has specifically excluded the surface hardening of steel, since such parts are predominantly magnetic at room temperature. However, we must not overlook the fact that in all surface hardening or through-hardening of steel, a portion of the part has been heated above the magnetic transformation point and must be considered "nonmagnetic" in any analysis of heating by induced currents.

Other things being equal, the higher the frequency the greater will be the tendency for the heating to be confined to the surface layers. However, where the heating time is of the order of several seconds or more, heat flowing inward by conduction accounts for most of the "depth" of temperature penetration.

For the heating of large numbers of parts per hour, the efficiency of conversion in terms of B.t.u. per kw-hr. determines the kilowatts required. High kilowatt requirements dictate the use of motor-generator frequencies, merely from the standpoint of economics. The cost of complete induction heating units developing up to 30-kw. output averages around $300 per kw., regardless of the type of generator used to produce the high-frequency current. At higher output levels, this factor remains the same for vacuum tube and spark gap equipment, but falls off rapidly for motor-generator sets, and will reach a value of as low as $50 per kw. at power output ratings in excess of 1000 kw.

Conclusion

This paper has been presented with the hope it may correct some of the misunderstandings which have been so apparent when metallurgists consider the use of high frequency for heating nonmagnetic materials. The conclusion of the analysis is that the heating efficiency is essentially independent of current frequency, provided the material has a sufficiently large cross section to respond to the normally used induction heating frequencies of 3000 cycles and above. Therefore, the selection of frequency depends upon the other factors fundamentally of an economical or practical nature.

Fig. 9 — Five-Station Fixture for Heating Aluminum Propeller Hubs for Forming Flanges. A 200-kw., 3000-cycle unit provides the power. A 9-in. length of the 5¼-in. diameter hub is heated to 850° F. at the rate of one every 47.5 sec.

Dynamic Hot Hardness Testing

(with special reference to isothermal transformations)

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To investigate some irregularities in hardening a die steel, a quick determination of its martensite start point was made by a series of punch marks during air cooling. The method used — measuring relative indentation hardness by spring-loaded center punch — seemed to have wider utility and some trials were made to determine the isothermal transformations (TTT-curves) of two steels with fairly high alloy content. Results on S.A.E. 52100 checked published S-curves and were verified by examination of microstructures.

NUMEROUS investigators have studied hardness testing at elevated temperatures. It is neither necessary nor desirable to review in detail their various methods, since the techniques generally differ from conventional methods at room temperature only in the use of special apparatus to maintain temperature and to minimize the effect of heat on the mechanisms employed.

It may be said that hardness tests are made by either *(a)* placing a steady load on a suitable indenter, as in Rockwell or Brinell tests, or *(b)* producing an indentation by a suddenly applied blow acting through a suitable penetrator. In the first method the penetrator must remain in contact with the hot metal for a considerable time, and in order to be sure of maintaining a definite temperature of the metal, special precautions must be taken. E. C. Bishop and M. Cohen have recently reported on these techniques. (See *Metal Progress* for March 1943, p. 413.)

A test method which would not incorporate parts of the hardness tester within the furnace would seem to have certain advantages. As is well known, there are portable hardness testers available, so designed that a blow is struck and a Brinell ball driven into the material being tested. Results are translated into conventional Brinell numbers by comparison of impressions made by the portable machine on blocks of standard hardness. In such a dynamic test the penetrator is not in contact with the material for any appreciable time.

Some years ago the senior author, at University of Cincinnati, studied the use of a spring-loaded center punch as a hardness tester. The ordinary prick punch, used by mechanics, was altered to use various designs of penetrators, such as a $\frac{1}{16}$-in. steel ball or a hardened steel cone. Reproducible results and good conversion factors could be found for evaluating hardness in terms of Rockwell B and C-scale, providing certain precautions were observed: The sample must be supported firmly; spring characteristics, in terms of energy released, must be maintained constant throughout the program; each commercial punch must be calibrated. Diameters of impressions were read to 0.001 in. with a shop microscope.

The original object of the present investigation was to determine the martensite start point in certain alloy steels by following the change of hardness during continuous cooling from the aus-

tenitizing temperature. It is obvious that not all steels could be investigated by such a method, particularly using small samples, but preliminary trials indicated interesting possibilities for the air hardening steels. Later the investigation was extended to the isothermal transformations of two alloy steels. In such studies the time and temperature at which transformation starts or ends is of interest rather than an exact hardness value. Consequently, the authors have not attempted to correlate the observed hot-hardness with any conventional hardness scale.

The Punch Hardness Tester

A spring-loaded center punch of a common commercial type was modified by adding a point designed after the Rockwell "Brale" penetrator, made from high speed steel, hardened and precision ground (Fig. 1). In use, the load must be applied vertically with axis of the punch perpendicular to the surface being tested. To accomplish this, a jig shown in Fig. 2 was constructed.

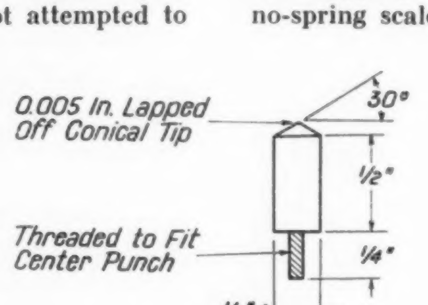
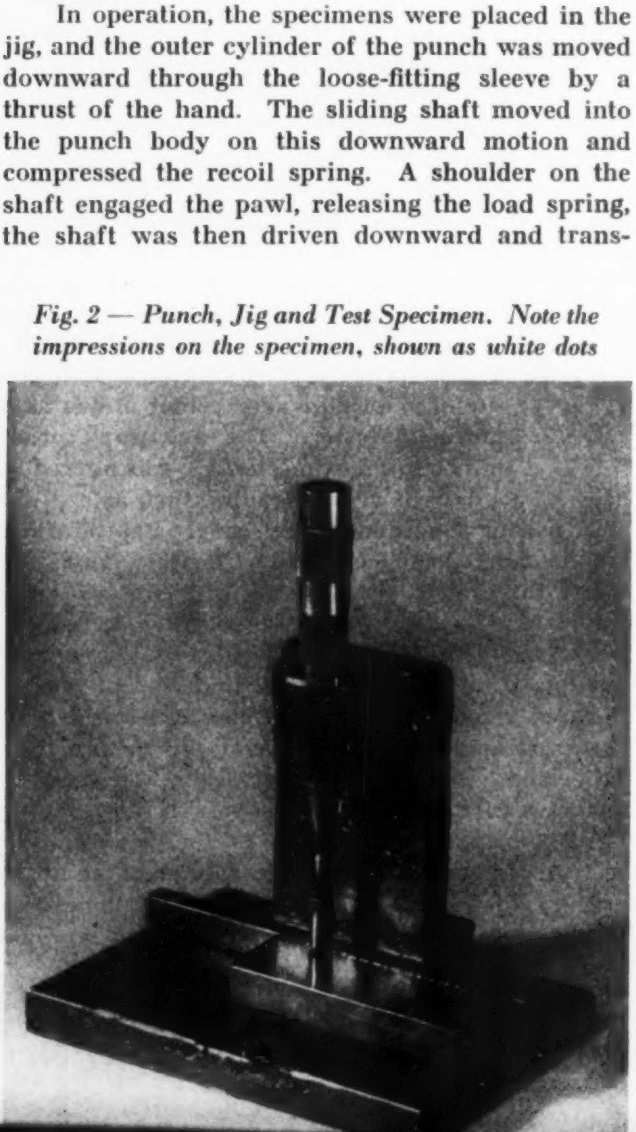


Fig. 1 — Detail of Punch Point

In operation, the specimens were placed in the jig, and the outer cylinder of the punch was moved downward through the loose-fitting sleeve by a thrust of the hand. The sliding shaft moved into the punch body on this downward motion and compressed the recoil spring. A shoulder on the shaft engaged the pawl, releasing the load spring, the shaft was then driven downward and transmitted the blow through the penetrator. Impressions were measured at exactly 100 diameters by projecting them on the ground glass screen of a standard metallograph. Three diameters were read and averaged.

Fig. 2 — Punch, Jig and Test Specimen. Note the impressions on the specimen, shown as white dots



It was desirable to make frequent checks on the constancy of the load necessary to trip the punch and the dimensions of the indenter point. The tripping load was measured by means of a no-spring scale. The point, as originally ground, was magnified 100 times and its contours outlined on a ground glass screen, and the point after use was checked by comparison. There was no change in the spring or contour during this investigation.

Steel samples were austenitized at the desired temperature for constant time while packed in cast iron chips. Our original experiments were designed to determine the martensite start point, and a thermocouple was inserted in the piece being tested and impressions made as it cooled in air. Impression diameters were plotted against temperature, with the result shown in Fig. 3. (The oil hardening steel was immersed during the test in a shallow oil bath, in order to be able to operate the punch, and the oil agitated.)

In the first attempts to determine the M_s point of an S.A.E. 52100 steel, the hardness began changing several hundred degrees above the expected point. The critical cooling rate had not been exceeded; therefore, the change in hardness was attributed to the formation of ferrite and carbide rather than martensite. This phenomenon suggested that changes in hardness as they actually occur during isothermal transformation could be determined. The following quotation from Davenport and Bain* is of interest:

> "Characteristic hardness changes result from transformation and, while somewhat complex, could be used to evaluate the progress of transformation. In this case indentations with a uniform pressure might be made in the specimen at intervals and their depth interpreted later."

Determination of S-Curves

The technique of determining TTT-diagrams (S-curves) by following changes in hot hardness involved the austenitizing and isothermal transformations of an oil hardening toolsteel (nominal

*"Transformation of Austenite at Constant Subcritical Temperature", by E. S. Davenport and E. C. Bain, *Transactions* A.I.M.E., Iron and Steel Division, 1930, p. 117.

analysis: 0.75% C, 1.00% Cr, 1.75% Ni) and S.A.E. 52100. The steel specimens, approximately 3x1x½ in., were austenitized at 1690 and 1615° F. respectively, and quenched in a molten salt bath maintained at a predetermined constant temperature. The salt was a eutectiferous mixture of sodium nitrate, sodium nitrite, and potassium nitrate which melted at 285° F. (This salt decomposes at about 1100° F. and was therefore unsuitable for high temperature use.) An electrically heated steel pot contained the molten salt and the jig, and also acted as an anvil for the hardness test. The surface of the test specimen was about ¼ in. below the surface of the bath. The melt remained transparent and the operation could be watched closely. Temperature changes were measured with a thermocouple and potentiometer.

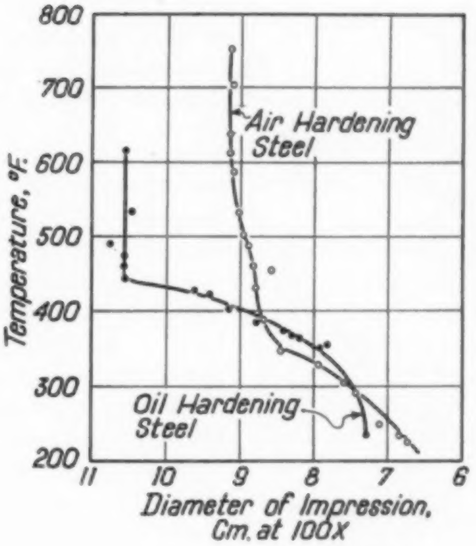


Fig. 3 — Change in Hardness of an Oil Hardening Steel (S.A.E. 52100) and an Air Hardening Die Steel (1.0% C, 5.25% Cr, 0.25% V, 1.1% Mo) During Martensite Transformation. Austenitizing temperatures were 1800° F. for both. M_s for 52100 steel is 440° F. and for the air hardening steel is 350° F.

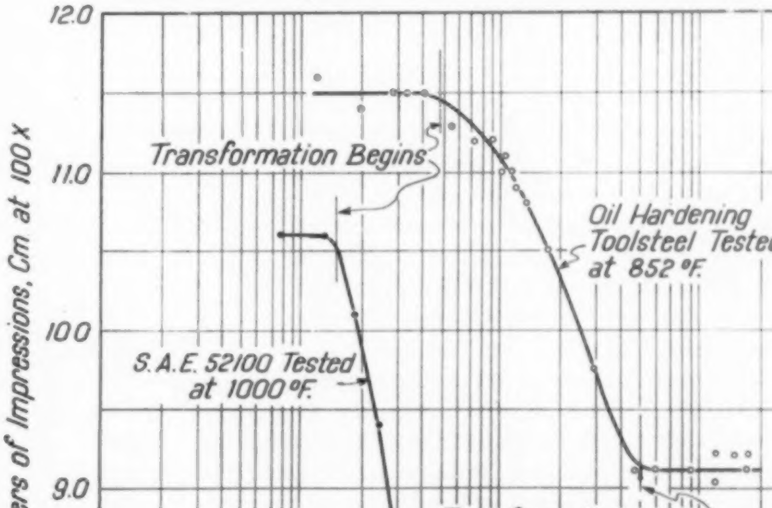
The salt bath used limited the investigation to temperatures below 1100° F. A few attempts to use lead for higher temperatures were unsatisfactory because of its buoyancy on jig and test specimen, as well as solidification of the lead on the penetrator. Some other salts investigated at temperatures above 1000° F. were unsuitable because of their corrosive effects. However, techniques can almost certainly be developed for temperatures higher than those here reported.

Once the steel had been quenched into the molten salt the center punch was inserted and a series of impressions were made. A minimum of 6 sec. was required to make the first impression; each additional impression took a minimum of about 6 sec. The punch was removed from the jig between impressions to prevent excessive heating.

Impressions were made at increasing intervals of time, since time was to be plotted on a logarithmic scale, the specimen being moved about ⅛ in. between impressions. The specimen was maintained at constant temperature and testing continued until transformation was complete or almost complete. The specimen was then water quenched and the indentations measured as described previously.

Curves were then plotted showing log time versus the impression diameter. Two examples are shown in Fig. 4. All curves so plotted had a horizontal portion for a certain length of time, depending on the isothermal temperature, and then a rather rapid break downward. If testing was continued until transformation was completed, the curves once again became horizontal. An arbitrary beginning and end of transformation was indicated on the diagrams, as shown in Fig. 4. When a sufficient number of points were obtained, the portions of the TTT-diagrams below 1000° F. were plotted for S.A.E. 52100 and the oil hardening toolsteel. These are shown in Fig. 5

Fig. 4 — Change in Hardness of the Two Steels During Isothermal Transformation at Temperatures Indicated



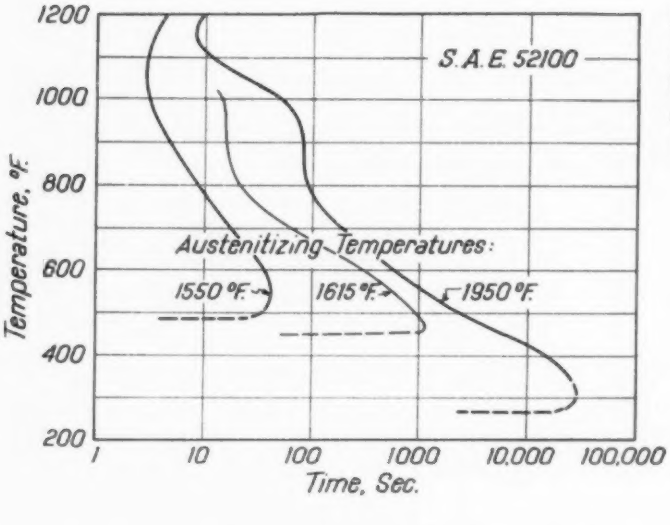


Fig. 5 — Beginning of Isothermal Transformation of S.A.E. 52100 (1.0% C, 1.4% Cr, 0.4% Mn, 0.3% Si, 0.2% Ni), Austenitized at Various Temperatures. Center curve determined by hot hardness tests; other two are from U.S. Steel's "Atlas of Isothermal Transformation Diagrams"

and 6. Figure 5 shows curves from two standard TTT-curves for comparison. For any temperature selected for study at least two series of hot hardness tests were made.

The validity of the transformation times as determined in this manner was checked satisfactorily by microscopic analysis of small specimens in a manner similar to that used by Davenport and Bain.

Discussion of Results

Figure 4 shows that the impressions become smaller during isothermal treatment, indicating the transformation of austenite to harder constituents. Microstructural studies gave a good check on the accuracy of transformation time so indicated.

Since it was possible that the chilling effect of the relatively cold penetrator might have some effect, tests were made with hot and cold penetrators and these tests indicated that the temperature of the penetrator had no significant effect.

Determination of martensite start points for the oil hardening toolsteel were not satisfactory, but for the S.A.E. 52100 steel our findings correlated very closely with published literature.

Consideration was given to the change in diameters of the impressions due to thermal contraction on cooling. The limits of accuracy in measuring diameters was about 3%, and the amount of change in length due to thermal expansion was less than 1%. Therefore, thermal contraction had negligible effect.

The hot hardness testing technique developed in this investigation gave a measure of the change in hardness or resistance to plastic deformation of the metal rather than a measure of any actual hardness value expressed in any of the conventional systems. During isothermal transformation the change in hardness of the steel from one hardness level to another was measured, but no numerical values of hardness were obtained other than the diameters of the impressions. It was desired to know at what *time* hardness changed rather than the *magnitude of change* in terms of usual hardness numbers. Transformation was frequently not carried to completion and the end of transformation was thus not obtained.

The method here discussed offers a rapid means of determining the isothermal transformation characteristics of oil hardening and air hardening steels below 1000° F. The method is not readily applicable to steels whose transformation begins in much less than 15 sec., although, as previously noted, an impression could be made in 6 sec. Accuracy, as compared with other methods, does not seem quite so great. However, it may be possible to determine the martensite start temperature for any oil hardening steel or air hardening steels, by this method, as well as predicting fairly closely

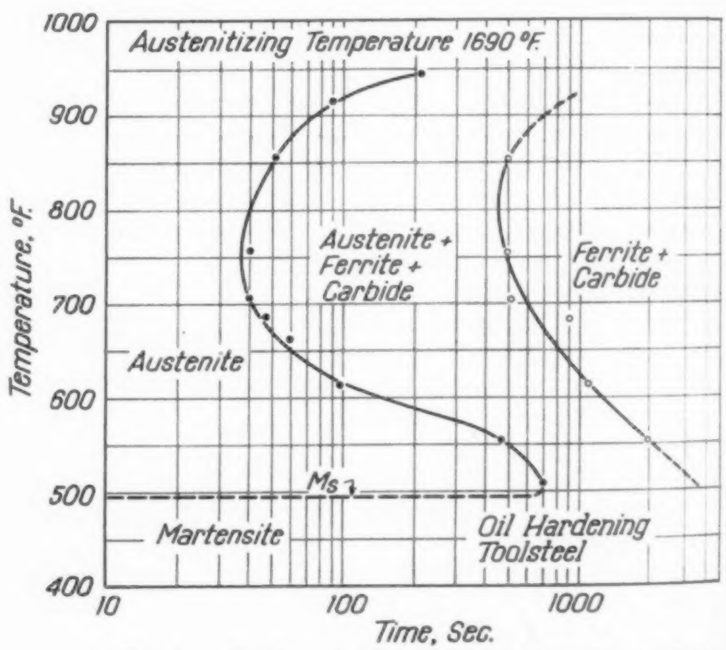


Fig. 6 — Portion of the Isothermal Transformation Diagram for the Oil Hardening Toolsteel (0.75% C, 1.0% Cr, 1.75% Ni) as Determined by Hot Hardness Tests

the beginning and end of an isothermal transformation.

Of the various methods used to determine isothermal transformation, microstructural changes give the direct measure of what is ultimately desired, and changes in hardness are closely related to changes in structure. Other methods measure changes in length or magnetic permeability and indirectly relate these to the changes in hardness and structure obtained during heat treating. The principal use of this technique would seem to be for preliminary explorations of the TTT-curves of a steel, or to determine its martensite start point. Since the location of the TTT-curves is often changed by a change of austenitizing temperature, a relatively quick check of possible effects can be made by the hot hardness method. This is demonstrated in Fig. 5.

This technique also has the advantage of simplicity, since no expensive equipment is needed. The usefulness of this method of hot hardness testing may, of course, extend to alloys besides steel, but so far no investigations except those reported here have been undertaken.

Penetrators other than the high speed steel point should be tried, and our future experiments will use carbide penetrators. However, the high speed point, as would be expected, stood up well in the temperature ranges investigated. There are obviously many improvements possible in technique, particularly in instrumentation for recording automatically the time and temperature when an impression is made.

Summary and Conclusions

A progress report on the use of a punch hardness tester for hot steel has been presented.

1. The equipment used and techniques employed are simple and time saving.

2. Exploratory experiments indicate reasonable accuracy in locating martensite start points in air hardening and in some oil hardening steels.

3. A hardness impression can be made as frequently as every 6 sec.

4. The isothermal transformations investigated so far have been limited to temperature ranges below 1100° F. Such curves as have been determined agree reasonably well with available data for the steels concerned.

5. The method seems to be of value in preliminary investigations of transformations, and particularly valuable where great accuracy is not needed.

6. Certain precautions are recommended in using the test. The exact value of blow struck is not important, but the blow must be reproducible, and hence spring characteristics must be checked frequently. The contour of the penetrator must also be checked to be sure that deformation of the penetrator has not occurred in testing.

7. Very light scale on the piece apparently does not interfere.

8. The use of a suitable jig is recommended, so that the punch may be maintained vertical during test.

Critical Points

By The Editor

TO the American Society for Testing Materials' meeting in Detroit and listened to some generous words concerning 20 years of officership on the Committee B-2 on Nonferrous Metals, and got to reminiscing on the changes time brings to institutions as well as men. Since the late great William Campbell of Columbia University resigned the chairmanship, the importance and diversity of specifications for the various cast and wrought copper alloys, the light metals, the die-cast metals, and the metal powders has grown so much that these portions of the work have been split off the old Committee B-2 and given a worthy independent status. More generally, the preponderance of interest in the A.S.T.M. seems to have shifted to nonmetallic subjects, to judge by the number of subcommittee meetings at Detroit devoted to cement, paints, petroleum, rubber — there being more than three times as many as meetings considering steels,

Shifts in interest in nonferrous metals

alloys, metallography, corrosion. This is a complete reversal of the proportion a generation ago. . . . One now seldom hears the oft-repeated fear that specifications tend to fix the status of a material, to check progress and to stifle improvement. How false this is — at least under the Society's procedure — is proven by the fact that two of the earliest specifications (for stranded copper conductor and slab zinc) were revised at this very meeting in Detroit. Zinc die castings are another notable example of continuous progress; minimum impact strengths are now more than double the maximum hoped for when the specification was first written. . . . As to formal procedure at the A.S.T.M. open meetings, a new member might say that they are affairs where decisions are railroaded through. He does not yet know how much work precedes these routine actions. Sometimes years are required to bring into common agreement two or three strong-minded groups of men, originally approaching a problem with diverse preconceptions and habits of thought stemming from long-established industrial customs. In every instance the first problem to be ascertained is "What are the facts?" Next: "What are the real needs?" Once these are mutually understood, the prejudices melt away. Slow though progress may appear at first in such a project, this retiring chairman of Committee B-2 has never seen one instance where agreement between producers and consumers has been impossible.

Finding the facts brings agreement

IT IS TOO BAD that this engineering method of finding a basis for mutual agreement cannot be extended into wider fields — into the arena of political action, for example. (Such was the common desire expressed by many notable speakers at the recent inauguration of T. Keith Glennan as the fourth president of Case Institute of Technology in Cleveland.) How well it would be if the needs and aspirations of a city-full of Americans or Russians could as accurately be determined as the properties and capabilities of a car-load of pig lead! A politician, or perhaps even a sociologist, might say that the human unit has such infinite diversity that accurate appraisal of its mass action is impossible. The engineer would insist on giving it an earnest try, but there is little likelihood that he will get a chance in the near future. Many were the reasons advanced by speakers at Dr. Glennan's inauguration for increasing the responsibilities of scientists and engineers in public affairs, and giving him a broader training therefor, but a cold-blooded analysis by Oliver E. Buckley, president of Bell Telephone Laboratories, concluded that engineering and politics simply do not mix. Many references were made throughout to the need of an awakening of the moral conscience and a revival in religious belief. Perhaps this is a step precedent to the plan advanced by David Lilienthal, chairman of the Atomic Energy Commission, who believes that every man or woman receiving a college education owes his country, in repayment, a few of the most vigorous years of his life in some form of public service.

Enlarged responsibilities of engineers

SUCH PIOUS HOPES for sane national management assume a sense of urgency when one listens to top army, navy and air officers discuss current planning, as they did at Selfridge Field during the June meeting of Army Ordnance Association. ("The unified strategic plan for America's defense is already formulated and we seek superiority of equipment rather than preponderance of manpower" — Chief of Staff Bradley. "We plan to have no munitions manufacture of importance closer than 200 miles of the coastline" — Chief of Army Ordnance Hughes. "The Navy's problem is to sweep the enemy from the sky, sea, and underwater, so we can safely transport an invasion force" — Chief of Naval Ordnance Noble. "One principal line of naval research is toward detection of high speed, deep cruising submarines, equipped with antisonic devices" — Assistant Chief of Naval Operations Momsen.) Both the still and the live exhibits at Selfridge Field emphasized the matter of speed, speed to the utmost. The speed and altitude of jet fighters and bombers — to say nothing of guided missles — brings problems of an entirely new scale of magnitude to anti-aircraft and fire control. No longer is it possible to detect and track an aircraft as it approaches the target and shoot at it after it passes; it goes away from you so fast the projectile has some difficulty catching up! Likewise since it may carry an atomic bomb, it must be intercepted *before* it arrives. Either for offense or defense there is no point in getting devices that will give "near misses", so we must perfect homing missles to intercept targets moving at supersonic speeds. . . . The whole display and the maneuvers, despite nerve-tingling moments, was most depressing. Without being able to suggest an alternative to Army Ordnance Associations' slogan: "Keep America Strong for Peace", a disillusioned survivor of two world wars could not help but remember two others, "A war to end wars" and "A war to make the world safe for democracy", which somehow have a hollow sound in 1948.

The problems of defense

Properties of Aluminum Bronzes at Subzero and High Temperatures

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A comprehensive series of tests on three typical aluminum bronzes, over a wide range of temperature, indicates maximum strength and elongation with only slightly impaired Charpy impact figures at −295° F. for the alpha bronze. Sharp drop in tensile strength, as well as Charpy impact, occurs in both alpha and duplex bronzes at about 600° F. Welding of wrought specimens decreases the impact and elongation values at subzero temperatures, as compared to unwelded metal under the same conditions.

THERE ARE TWO main reasons why data on the low-temperature properties of metals are of current technical interest. One is for design information on materials of construction for gas liquefaction equipment, which operates at very low temperatures. The other is the need for materials for service in extremely cold climates. There is also need for additional information on high-temperature properties of aluminum bronzes because many of the applications are based on their excellent oxidation resistance.

Because of these factors, an investigation was undertaken at Battelle Memorial Institute under sponsorship of Ampco Metal, Inc., of Milwaukee, on the effect of low and high temperatures on the mechanical properties of three aluminum-bronze alloys in both the wrought and as-welded conditions. It resulted in data on tensile properties (ultimate, yield stress at 0.5% strain, and percentage elongation), Brinell hardness, Charpy impact, and dilation (linear thermal expansion). Testing temperatures were −295, −75, −20, room temperature, 400, 600, 800, and 1000° F.

Previous Work — Concerning the properties of aluminum bronze at low temperatures, only the data reported in 1933 by E. W. Colbeck and W. E. MacGillivray in *Transactions* of the Institution of Chemical Engineers are available. These investigators found, on lowering the temperature of an alpha alloy (7.31% Al, 1.02% Zn, 0.44% Mn, 0.056% Fe, 0.018% P, balance Cu) from ambient to −292° F., that the tensile strength increased from 77,300 to 96,100 psi., the 0.1%-offset yield strength increased from 26,700 to 29,200 psi., the elongation in 2 in. increased from 26 to 28%, and the Izod impact value decreased from 24 to 20 ft-lb. This indicated a very desirable set of circumstances, namely, strength and ductility increase with slight decrease in impact properties at subzero temperature.

Concerning short-time tensile tests at elevated temperatures, Walter Rosenhain in 1907 reported in the *Proceedings* of the Institution of Mechanical Engineers some work on binary Cu-Al alloys (rolled bars) containing 6.73% Al and 9.9% Al. Three years later, Rosenhain and Lantsberry described three ternary Cu-Al-Mn alloys in the same *Proceedings*. These were cold drawn bars containing *(a)* 9.9% Al, 1.0% Mn; *(b)* 10% Al, 2.0% Mn; and *(c)* 9.1% Al, 2.8% Mn. Most

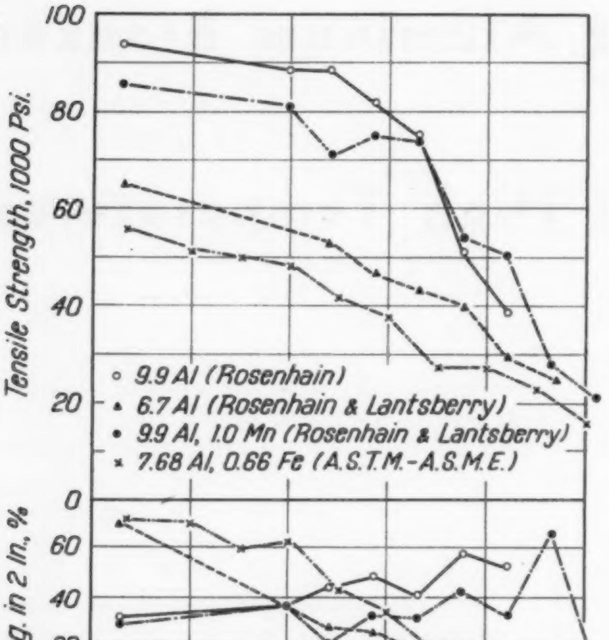


Fig. 1 — Some Published Data on Short-Time, High-Temperature Tensile Properties of Four Aluminum Bronzes

recently, in 1938, the joint A.S.T.M.-A.S.M.E. Committee on Effect of Temperature on Properties of Metals reported work on a ternary Cu-Al-Fe alloy (die castings) containing 7.68% Al, 0.68% Fe, 0.08% Sn, 0.01% P.

Some selected results of these investigators are given in Fig. 1. It is seen that tensile strength of the alloys decreases progressively with rising temperatures — somewhat more rapidly at temperatures above 600 °F. for the wrought duplex alloys of Rosenhain than for the alpha Cu-Al alloys. Although the ductility (elongation) of the alpha bronze with 6.7% Al is much higher than the duplex bronzes at room temperature, it decreases with increasing temperature, whereas the elongation of the duplex bronzes remains at about the same level up to 850° F.

Materials

All test specimens were made and machined into test specimens at the Ampco plant in Milwaukee from 2-in. extruded bar stock. Typical analyses of the three alloys investigated are listed in Table I.

Alloy Grade 8 is a ternary aluminum bronze of the alpha type having iron present as a grain stabilizer and also for the purpose of improving physical properties. The microstructure for this alloy, as annealed at 1150° F. and quickly cooled, is shown in Fig. 2. It consists of twinned alpha grains, with a grain size of about 0.025 mm. The principal application for this alloy is as a material of construction for the chemical industry, usually sheets, tubes, and plates for tanks, housings, heat exchangers, agitators and condensers.

Fig. 2; 250 × Fig. 3; 250 × Fig. 4; 100 ×

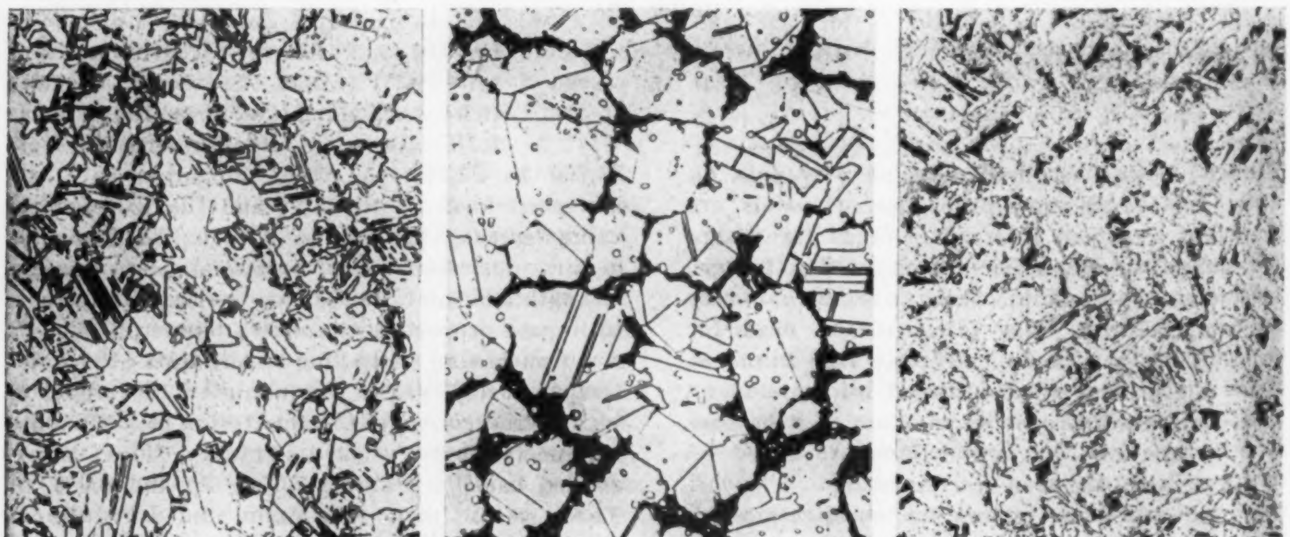


Fig. 2 to 4 — Alloys 8, 15 and 45, Respectively, Annealed at 1150° F. and Quickly Cooled; Etched With 0.5% Ferric Chloride

Alloy **Grade 15** is a ternary aluminum bronze of the duplex type, having iron present for strength and grain stabilization. The microstructure of the alloy, as annealed at 1150° F. and quickly cooled, is shown in Fig. 3. It consists predominantly of twinned alpha grains with some acicular beta (retained by quick cooling) outlining the alpha grains, and isolated round particles of iron-aluminum constituent dispersed throughout. It is this intermetallic compound, with its high hardness and dispersal throughout the structure, which is responsible for the alloy's excellent wear properties. It is used chiefly for moving parts subjected to wear and friction, such as gears, shafts, bushings, valves and valve seats.

Alloy **Grade 45** is an aluminum bronze of high strength having additions of both iron and nickel as well as a little manganese. As may be seen in the microstructure in Fig. 4, the annealed alloy is of the duplex type, with the alpha-beta arrangement characteristic of slow cooling from the single-phase beta field into the alpha-beta field — a structure similar to the well-known "muntz metal" structure. There are tiny particles of iron-nickel-aluminum constituent distributed throughout. This alloy is used chiefly for structural parts subjected to static loads and not too much movement, such as studs, bolts, nuts, and valve seats.

Alloy **Grade 45-22** is a heat treated version of the Grade 45. The treatment constitutes a high-temperature quench from 1650° F. into water, followed by aging at 1200° F. The microstructure, as seen in Fig. 5, consists of very large grains of beta, with needles of reprecipitated alpha distributed throughout the grains or emanating from the boundaries. The iron-nickel-aluminum constituent is still present, although not visible in the dark background.

Table II lists the mechanical and thermal treatments for the four alloys as they were tested. As may be seen, the cold finished condition differs from the annealed condition by only a small amount of cold work. The rate of cooling from the 1150° F. anneal is fast enough to retain the beta phase in its acicular form, and prevent any decomposition into eutectic.

Table I — Typical Chemical Analysis of Material

Alloy and Description	Cu	Al	Fe	Ni	Mn	Si	Zn
Alloy 8, composition	89.62	7.78	2.57	0.01	—	—	—
Weld rod composition for welding	89.00	10.00	0.90	0.10	—	—	—
Nominal composition of weld metal	90.06	8.04	1.35	0.09	0.02	0.42	—
Alloy 15, composition	87.20	9.20	3.10	0.40	0.02	0.02	0.05
Weld rod composition for welding	84.80	11.30	3.50	0.40	0.02	—	—
Nominal composition of weld metal	86.00	9.40	3.75	0.35	0.02	0.42	—
Alloy 45, composition	81.20	10.10	3.00	4.75	0.80	0.02	0.05
Weld rod composition for welding	81.20	10.10	3.00	4.75	0.80	0.02	0.05
Nominal composition of weld metal	82.40	8.75	3.00	4.60	0.75	0.50	—

Fig. 5; 100 ×

Fig. 5 — Alloy 45-22, Quenched From 1650°, Aged at 1200° F. 100×. Etched with 10 g. $FeCl_3$, 100 ml. HCl, 100 ml. H_2O

Welding used the carbon-arc method. Typical compositions of weld rods and metal in weld joints are listed in Table I. The manner in which the welds were made and the section of the bar from which the specimens were cut are shown in Fig. 6 and 7.

Testing Methods

Tensile tests at elevated temperatures used standard equipment and followed the procedures outlined by A.S.T.M. Specification E21-43. At subzero temperatures, essentially the same equipment was used, except the specimen was surrounded by coolant in an insulated tank instead of a furnace.

Brinell hardness tests at the various temperatures were also fairly straightforward, using an indenter equipped with a 10-mm. Carboloy ball. The hardness specimen was a ½-in. transverse slice of the 2-in. extruded bar, prepared as shown in Fig. 7. Precautions were taken to heat or cool the penetrator along with

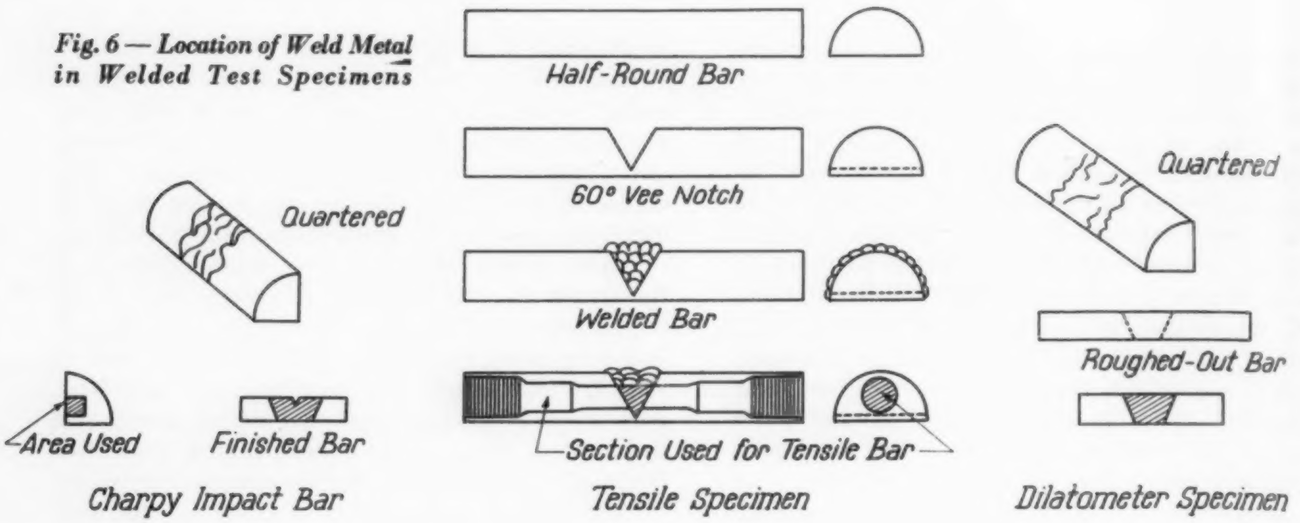


Fig. 6 — Location of Weld Metal in Welded Test Specimens

Table II — Mechanical and Thermal Treatment of Specimens

Condition	Operation	Alloy Grades			
		8	15	45	45-22
Cold finished	1. 2-in. bar extruded	x	x	x	x
	2. Annealed at 1150° F., quickly cooled	x	x	x	x
	3. Cold drawn, in.	⅜	$\frac{1}{16}$	$\frac{1}{16}$	$\frac{1}{16}$
	4. Annealed 1150° F., quickly cooled	x	x	x	x
	5. Roller straightened, in. cold reduction	0.003	0.003	0.002	0.002
Annealed	6a. Annealed 1150° F., quickly cooled	x	x	x	x
Welded	6b. Welded in the carbon arc, and machined	x	x	x	x
Heat treated	6c. Water quenched from 1650° F.★				x
	7. Aged at 1200° F., air cooled				x

★1 hr. at temperature per in. diameter of bar.

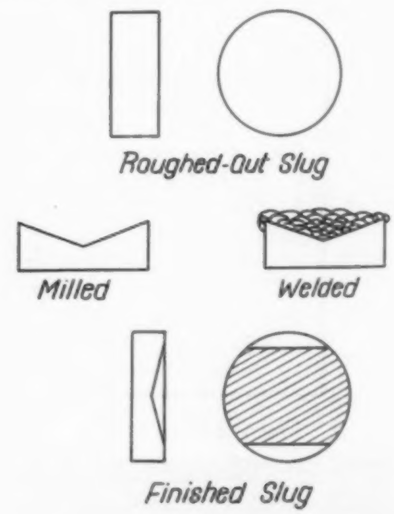


Fig. 7 — Method of Preparing Slug for Measuring Hardness of Weld Metal

the specimen before testing. Loads were 3000 kg. at temperatures from −295 through 600° F., and 500 kg. at 800 and 1000° F.

Charpy impact tests were made on standard V-notch bars (Fig. 6) having dimensions in accordance with A.S.T.M. Specification E23-41T. An initial energy of 58 ft-lb. on an Amsler machine was used for all tests except those on alloy Grade 8, which required an initial pendulum setting of 93 ft-lb. Specimens were held at temperature for at least 15 min. before insertion into the testing machine.

Dilation experiments were run on specimens 3 in. long and 0.5-in. diameter, in the temperature range −150 to 1200° F., according to the procedures outlined for the "rapid method" given by A.S.T.M. Specification B95-30.

At least three separate tensile or impact tests were run for an alloy at a given temperature and the values were averaged. Three dilation runs were carried out and an average dilation curve was determined for each of the alloys in each of the conditions tested, from which mean linear thermal expansion coefficients were calculated.

Two specimens with six impressions per specimen were used for each hardness determination. Subzero temperatures were obtained with the use of liquid oxygen for −295° F. and mixtures of dry ice and acetone for −20 and −75° F.

Results of Tests

Mechanical Properties — Results of tensile, hardness, and impact tests on the three alloys investigated are shown in Fig. 8 and 9. As would be expected, the tensile strength, yield strength,

and hardness generally increase at subzero and decrease at elevated temperatures. Above 600° F., the decrease in the strength properties and hardness is particularly marked; coincident with the drop-off in tensile properties is an increase in elongation, which shows a maximum at 800° F. Ductility generally increases in these wrought alloys with decreasing temperature below 600° F., reaching a maximum at liquid air temperature. The exceptions to this are the various forms of Alloy 45, which drop in elongation from −75° F. Welded test specimens invariably showed poorer elongations at subzero temperatures.

There is little difference between the properties of alpha alloy, Grade 8, in cold finished and annealed conditions. At −295° F. both show moderately high tensile and yield strengths, very high elongations, and high impact values (Fig. 8). The welded condition differs from the wrought metal in lower elongation at subzero temperatures,

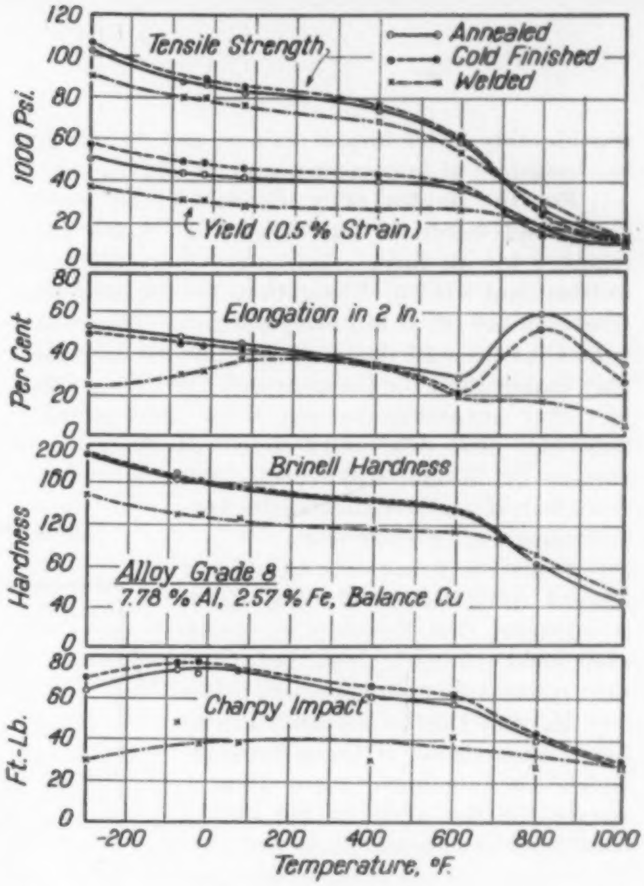
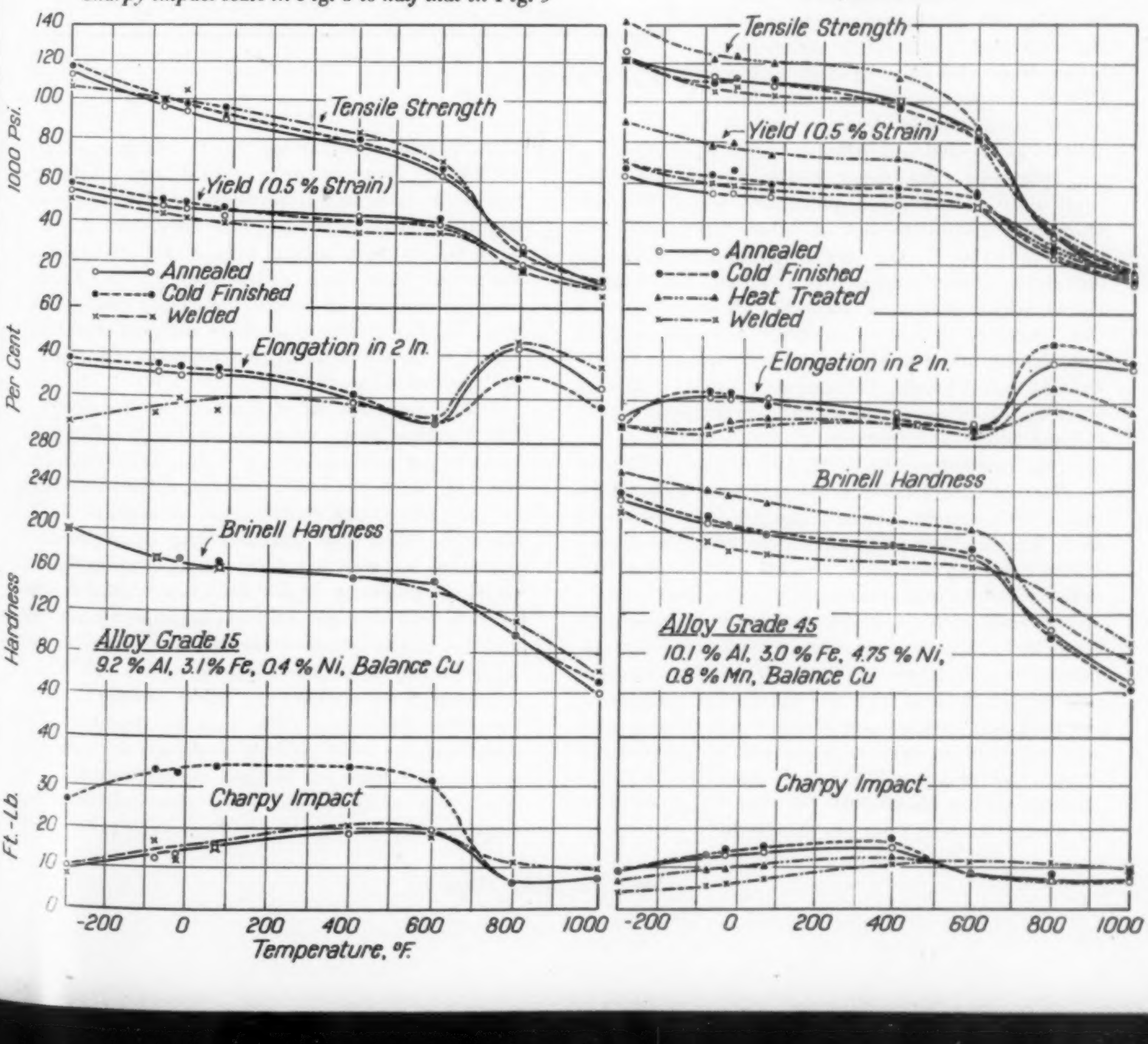


Fig. 8 (Right) and Fig. 9 (Below) — Mechanical Properties of Alloys Grade 8, 15 and 45 as They Vary From Liquid Air to 1000° F. Note that Charpy impact scale in Fig. 8 is half that in Fig. 9



considerably lower impact values, and no increase in elongation at temperatures above 600° F.

For the duplex alloy, Grade 15, the tensile, yield, and hardness curves in Fig. 9 are close together for all three conditions — annealed, cold finished and welded. Elongations for the as-welded condition fall off at low temperatures. The impact values for the cold finished condition are considerably higher than for the annealed condition, which is rather surprising inasmuch as there actually was very little cold work done. Rechecks were run on additional specimens cut from bar stock, but the results corroborated the original tests.

From the results for the highly alloyed Alloy Grade 45 (Fig. 9), it is apparent that the gain in tensile and yield strength resulting from heat treatment is effective only below 600° F., and this is associated with lower elongations at these temperatures. The tensile and yield strength curves for the alloy in the duplex condition, as annealed, cold finished, or welded, are all comparable each to each, but the hardness curve for the as-welded condition is somewhat lower than the others when below 600° F. Impact values are fairly low at all temperatures, but are lowest for the as-welded condition at −295° F.

Dilation — The three dilation runs made on each alloy in the three conditions tested did not show any discontinuities at subzero temperatures down to −150° F., nor in the neighborhood of the eutectoid transformation — which for binary Cu-Al alloys occurs at 1060° F. The reason for this probably is that the acicular beta phase does not change into the alpha and delta phases representing equilibrium in the short time required to heat the alloy to the eutectoid temperature at normal rates (3° F. per min.).

H. Carpenter and J. M. Robertson in Vol. II of "Metals" describe this sluggishness with which the acicular beta phase changes into eutectoid upon heating, and therefore agree with the reason given here for the lack of any dilatometric evidence of transformation.

Linear coefficients of expansion as measured from 32° F. to various subzero and elevated temperatures are given in Table III, from which it appears that there are no significant differences between the various conditions for any one alloy nor — as far as that is concerned — between the alloys themselves. The coefficient of expansion increases from about 8.6×10^{-6} per °F. at −150° F. to about 10.7×10^{-6} per °F. at 1100° F.

Table III — Mean Linear Coefficients of Thermal Expansion* From 32° F., per °F.

Alloy	Condition	−150° F.	−60° F.	300° F.	570° F.	1100° F.
8	Cold finished	8.5	8.9	9.3	9.8	10.7
	Annealed	8.6	8.8	9.3	9.6	10.6
	Welded	8.7	8.9	9.5	9.8	10.7
15	Cold finished	8.9	9.1	9.2	9.6	10.9
	Annealed	8.9	9.0	9.0	9.5	10.7
	Welded	8.8	8.9	9.6	10.0	11.1
45	Cold finished	8.4	8.4	8.9	9.6	10.7
	Annealed	8.9	9.0	9.2	9.3	10.3
	Welded	8.4	8.9	9.2	9.5	10.7
	Heat treated	8.4	8.6	9.2	9.6	10.7

*Average per $°F. \times 10^{-6}$ from 32° F. to temperature shown at column headings.

Discussion

Our tests described above give no conclusive evidence of any structural changes taking place in the alloys as a result of the low-temperature treatment. To determine whether a low-temperature treatment would have any effect on subsequent properties or microstructure at room temperature, small sections of annealed Grades 8, 15, and 45 alloys and heat treated alloy 45-22 were tested for Vickers hardness (10-kg. load) before and after a 30-min. immersion in liquid air. After the liquid-air treatment, hardness values were very slightly less than before, but only 1 to 3 points. Also, micro-examination, before and after, did not show any significant differences. It is therefore believed that no additional structural change takes place in any of the aluminum bronzes tested as a result of low-temperature treatment (at least in an irreversible way).

The increase in elongation of the alloys at 800° F. takes place concurrently with a rapid loss in strength and hardness. Since the alloys have not been cold worked appreciably, this weakening is not associated with recrystallization. Also, since the effect was observed with the alpha alloy, Grade 8, as well as with those containing beta phase, it probably is not associated with the beta to eutectoid change. No explanation is available for this phenomenon other than a direct effect of temperature on atomic bonding.

Designers should use yield strength at elevated temperatures with caution, because creep data are the only reliable criterion for performance under such conditions. Continued deformation with time would take place at stresses considerably lower than the reported yield

strengths at 0.5% strain. (Use for design purposes of the data given for subzero temperatures is perfectly straightforward.)

Although creep tests were not performed in the present work, some creep data are available on one of the alloys covered by this article. J. F. Klement, chief metallurgist of Ampco Metal, Inc., who cooperated whole-heartedly throughout this work, reports that annealed Grade 15 alloy has a creep rate of 0.1% in 1000 hr. at 20,000 psi. and 500° F. Creep data have also been reported by the A.S.T.M.-A.S.M.E. Joint Research Committee for the die-cast aluminum bronze whose short-time tensile properties are given in Fig. 1: At 500° F. and 10,000 psi. applied load, this alloy containing 7.68% Al and 0.66% Fe showed a creep rate of 0.54% in 1000 hr.; at 800° F. and 3000 psi., the creep rate was 0.04% in 1000 hr.; and at 900° F. and 3000 psi., the creep rate was 0.35% in 1000 hr.

J. J. Kanter also reported in the same publication that a die-cast aluminum bronze of 7.5% Al showed the following creep properties: At 563° F. and 10,000 psi., the creep rate was 1% in 10,000 hr.; at 842° F. and 3000 psi., the creep rate was 1% in 10,000 hr.

Summary

The effect of temperature on mechanical properties between −295 and 1000° F. has been investigated for three representative aluminum-bronze alloys in a variety of physical conditions. Linear coefficients of thermal expansion for these alloys were also measured at subzero and elevated temperatures.

It was found that strength was highest at subzero temperatures and dropped off rapidly above 600° F. Accompanied with this drop in strength was a marked increase in elongation. Impact values were found to be highest for alloy Grade 8, which had an alpha structure. Elongation was found to increase as the temperature dropped from +70 to −295° F. for both the alpha Grade 8 alloy and the duplex Grade 15 alloy, and to decrease for the duplex alloy Grade 45 and beta alloy Grade 45-22.

Welding principally decreased the impact values and lowered the ductility (elongation) at subzero temperatures, as compared with the alloys in the wrought, unwelded condition.

Bits and Pieces

Any book you may choose from the A.S.M. publications (except Handbook) for an acceptable item for these pages.

An Improvement in Lead Laps

METALLOGRAPHIC polishing with abrasive embedded in a smooth lead plate (lead lap) instead of on a fabric covered steel disk, has certain advantages and is widely used. We have encountered some difficulties with the serrated cast iron plate used to work the abrasive into the lead lap, since corrosion products also became so embedded. Such foreign particles, of different size and hardness from the abrasive charged, generally played havoc with metallographic polishing. To overcome these difficulties the International Nickel Co. recommended "Type 2B Ni-Resist" as having the desired combination of high hardness and wear and corrosion resistance.

Several castings of this type were made and a pattern of diamond-shaped areas was machined in the smooth top surface by parallel grooves cut to a depth of $\frac{1}{16}$ in. The surface was then ground, and the edges of the diamond-shaped areas were filed and stoned free from burrs.

The surfaces of these castings have been repeatedly charged with abrasives mixed with water and the laps have been pressed and rotated on the surface to embed the abrasive in the lead. Six months' use shows the material to be eminently satisfactory from the standpoint of high resistance to wear and freedom from corrosion. If the charging disks remain covered in storage they need only to be wetted down to loosen the dried abrasive and then are ready for impregnating the lead laps. (WALTER H. BRUCKNER, research assistant professor of metallurgical engineering, University of Illinois)

Rapid Shop Test for Zinc Die Casting Alloys

ZINC base die castings or ingots, usually containing up to 3.5% copper and 4.5% aluminum, can be easily and rapidly distinguished from commercial and high purity zinc in the following manner:

Chip off a very small piece or use a small turning. Clean and put it in the depression of a white spot plate. Add to it 4 drops of a 50% nitric acid solution. If it is the copper-aluminum-zinc alloy, a dark gray-black suspension will stream out from the sample and color the liquid. If it is substantially pure zinc, the liquid will remain clear during and after the reaction. (Incidentally, the test serves also to distinguish zinc from aluminum and its alloys, since the latter do not react with nitric acid.)

The test can be applied directly upon the metal; however, it is sometimes difficult to prevent the drops from spreading and so diluting the desired effect. (C. GOLDBERG, chemist-metallurgist, New England Smelting Works)

Multiple Tempering of Low-Alloy Toolsteel

WHILE high-alloy toolsteels are commonly multiple tempered to transform the austenite retained after quenching — and the underlying theory is fairly well known — the practice is seldom applied to low-alloy steels, doubtless because it is thought that practically all their austenite transforms into hard martensite on quenching. However, multiple tempering solved a heat treating problem on Carpenter Steel Co.'s "Solar" water hardening toolsteel (0.50% C, 0.40% Mn, 1.00% Si, 0.50% Mo).

The problem arose out of the necessity of producing a screwdriver bit (minimum cross section 0.032x0.250 in.) able to withstand a twisting torque of 40 in-lb. By following the recommended heat treating instructions, the best result was 35 in-lb. Multiple tempering produced specimens that failed at torque values of 48 to 61 in-lb. In addition, the screwdriver bits, so treated, displayed unusual ductility considering that their hardness measured C-53 to 54.

The heat treatment schedule was: Heat to 1600° F., quench in oil (a 50° higher heat and a much slower quench than recommended by the manufacturer), temper 2 hr. at 200° F., air cool, follow with four other tempering heats for 1 hr. at 300, 400, 500, and 600° F. in succession, air cooling to room temperature after each. (B. Z. BERMAN, The Bermack Co.)

Resonant Frequency as a Means of Inspection

THE traditional method of testing a completed metal part or subassembly by striking it with a hammer and listening if it "rings true", can now be readily refined by using tachometer, coil magnet or oscilloscope to measure the frequency of a part which can be vibrated. It is merely necessary to clamp one end of such a piece in a vise and direct an air blast at the other, thus causing it to vibrate at its natural frequency. As expected, if one of several supposedly identical pieces vibrates at a lower frequency than the others, it is quite likely to be defective. Samples found to be faulty by this means have failed prematurely in the fatigue test, thus corroborating the existence of defects.

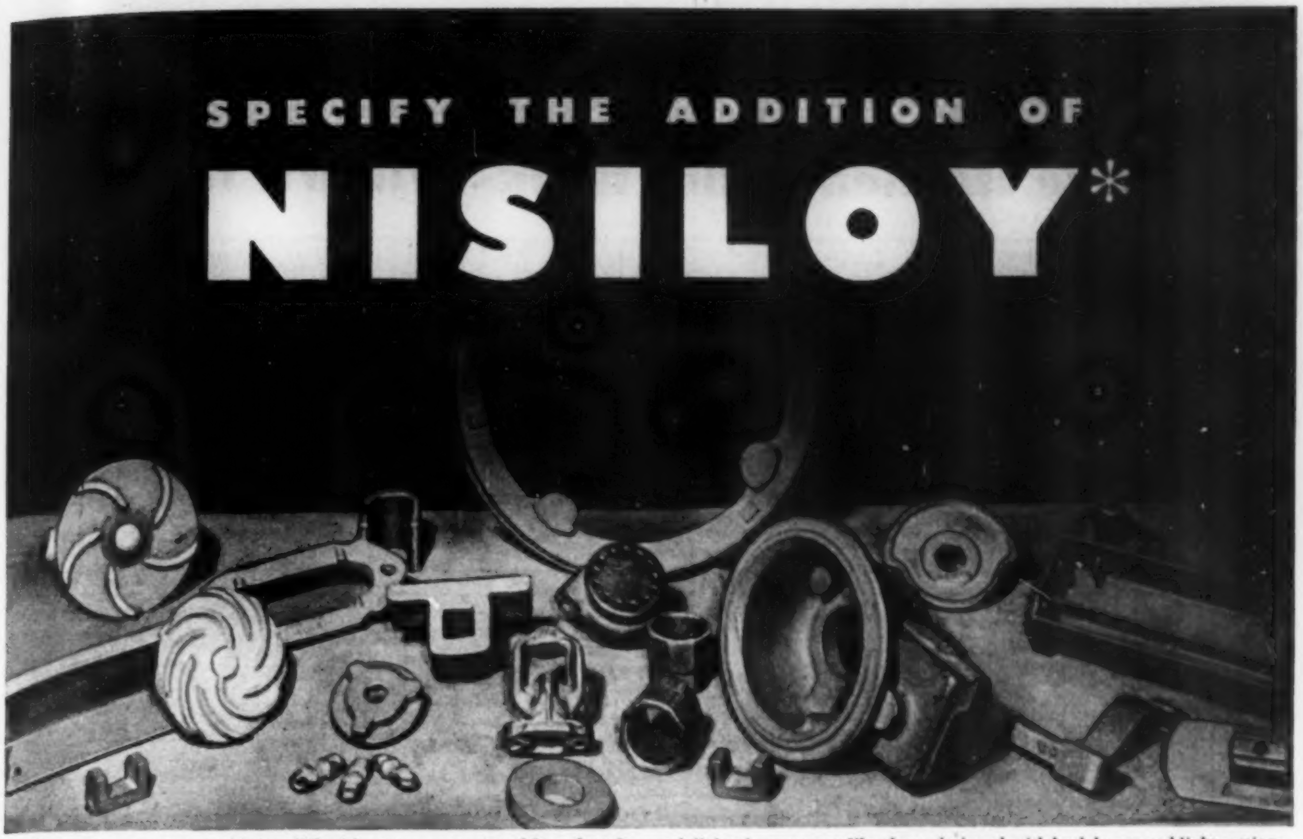
Of course such a resonant frequency test is nondestructive. Many times it may improve the original strength of the part, since mild understressing, as occurs during the vibrations, usually increases the endurance limit of the part. (EDWARD EPREMIAN, Research Laboratory, General Electric Co.)

Aluminum Spray Coating to Protect Welded Units

WHEN it is necessary to protect a pressure vessel from rusting, either in storage or after hydrostatic testing, it is often necessary to apply the protection *after* welding fabrication because of necessary stress annealing heat treatments. When openings are small or of complicated design this often prevents the use of many of the good plastic protective coatings.

We have found that if machined or grit blasted metal parts (plates, flanges, nozzles, which are to be joined by welding) are sprayed with a coating of metallic aluminum, about 0.003 to 0.005 in. thick, we will secure satisfactory protection against moisture corrosion. This coating is not oxidized by heating 24 hr. up to 1000° F., and will therefore survive stress relief heat treatments. It is also very adherent to properly grit blasted steel surfaces, thus insuring against loose particles which would damage metering equipment. The aluminum coating, when properly applied, will stand considerable plastic deformation of the steel parts without failure of the mechanical bond.

Our experience with aluminum sprayed metal also indicates it to be very satisfactory for outdoor exposure. (MERRIL A. SCHEIL, director of metallurgical research, A. O. Smith Corp.)



Chilling and consequent machining difficulties were encountered by a foundry specializing in cast parts like these, designed with both heavy and light sections. Nisiloy, added to the ladle, assured ready machinability after many other experiments failed.

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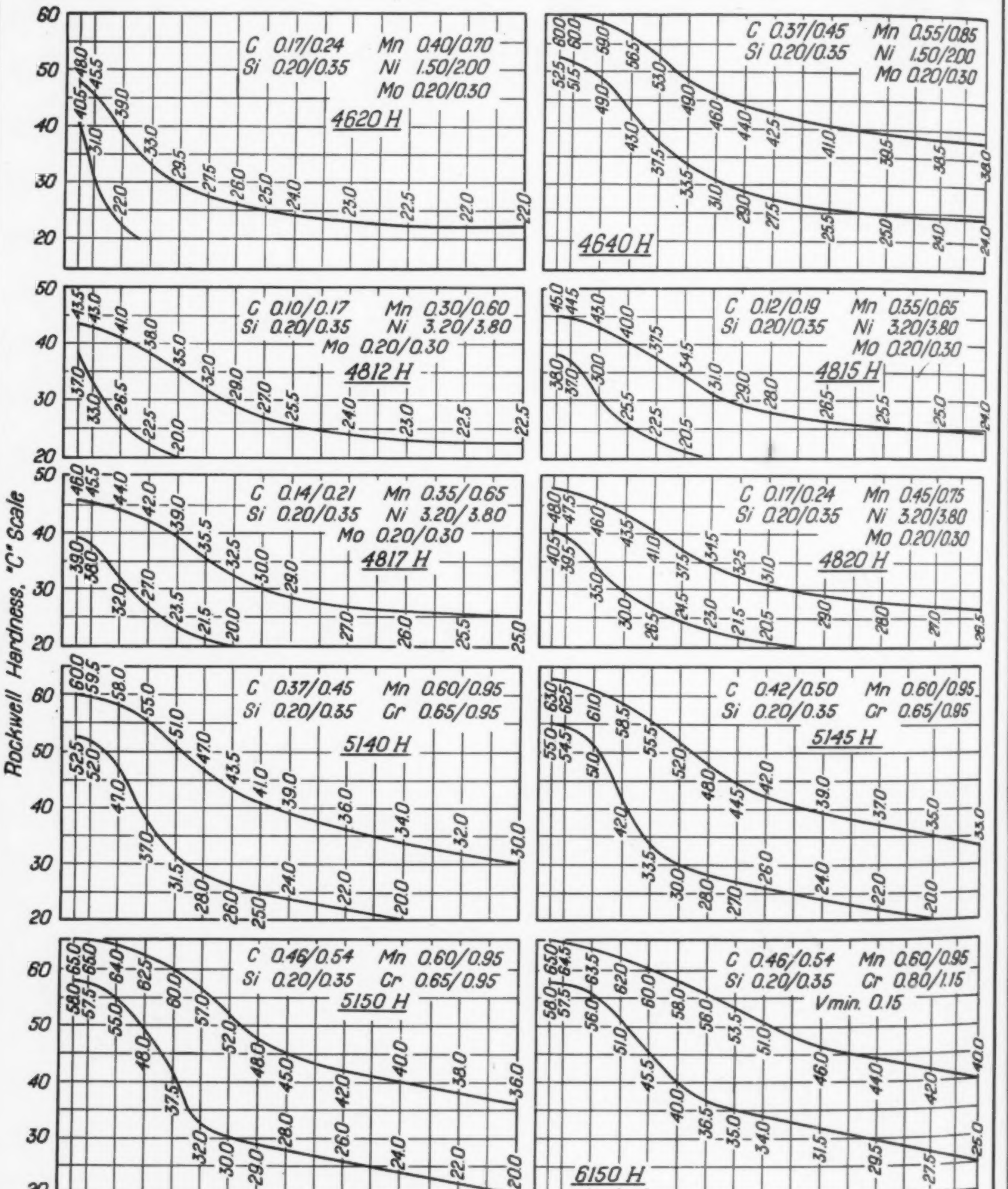


Galling Tests of Graphitic and Regular Oil Hardening Die Steels

By A. F. Sprankle
Timken Roller Bearing Co.
and R. W. Dayton
Battelle Memorial Institute★

Amsler wear testing machine was used to compare the antigalling characteristics of a conventional die steel of oil hardening, nondeforming type with a wrought graphitic steel (Graph-Mo), both hardened and tempered to C-61 or harder, when rubbing against soft S.A.E. 1015 steel of Brinell 90. Pressures were gradually stepped up to about 100,000 psi. If galling did not occur at these pressures at a low rubbing speed, the series was repeated at higher speed.

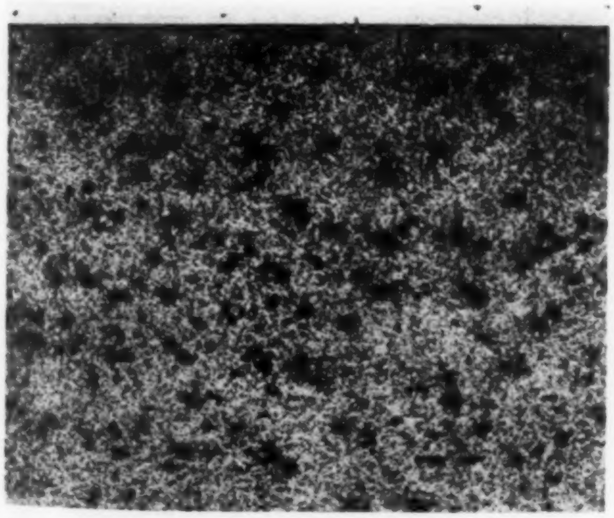


Fig. 1 — Structure of "Graph-Mo" at 100×, a Wrought Graphitic Steel, as Hardened. Constituents are tempered martensite, carbide particles, and graphite

UNINTERRUPTED PERFORMANCE of dies during cold blanking and forming operations is of paramount importance if maximum production is to be realized. There are, of course, many difficulties in achieving peak productivity, but by far the most troublesome technically is galling. This is the action that occurs when particles of the soft metal undergoing plastic deformation seize the hardened die that forms it. When this happens the die must be re-dressed, since scored or otherwise defective parts invariably result.

The usefulness of externally applied lubricants (that provide a nonmetallic film to combat this condition) has long been established and these are used in widely diversified forms. The possibility of using a lubricant inherent in the die material to augment the externally applied lubricant has not been so universally recognized. However, the excellent wear resistance of gray cast iron has been attributed by many investigators to the nonseizing qualities imparted by the graphite particles contained in the microstructure. With the inception of wrought graphitic steels, the desirable properties conferred by graphite in cast materials have been supplemented by the improved soundness and grain refinement of forged and rolled products. The micrograph of such material marketed under the name "Graph-Mo", represents the average structure of a hardened die; it consists of tempered martensite, carbides and graphite particles. The function of

★Of the authors, Mr. Sprankle is metallurgical engineer in the Steel and Tube Division of the Timken Roller Bearing Co., Canton, Ohio, and Mr. Dayton is assistant supervisor at Battelle Memorial Institute, Columbus, Ohio.

the latter will be discussed in some detail later on in this article.

Numerous tests have been made in actual production operations of wrought graphitic steels in cold blanking and forming operations, and their merit is quite well established. However, a direct quantitative comparison with a similar type of oil hardening, nondeforming, die steel has been lacking. The laboratory experiments to be described were undertaken to establish such a comparison.

A review of testing equipment available that would simulate most closely the service conditions of forming dies indicated that the Amsler wear testing machine would be the most suitable. The important considerations of this type of service are that the rubbing speeds are low, the unit loads are high, and the ambient temperature is in the vicinity of 70° F.

Materials and Heat Treatment

The test specimens used in the Amsler wear testing machine were 2 in. in diameter and 0.4 in. wide. All samples were machined 0.020 in. oversize on all dimensions, heat treated, and carefully ground and lapped to size.

Data on analyses and heat treatment are given in Table I. The S.A.E. 1015 was not heat treated since it was intended to represent the hot rolled strip steel that ordinarily would be processed by stamping or cold forming. The analysis of the oil hardening die steel was selected as one typical of those broadly classified as the nondeforming type.

Experimental Work — The test specimens were mounted in the Amsler machine so the die steel under test was rotated against a disk of the 1015 steel. These disks were mounted in the machine so that they contacted on their peripheries, as shown by arrows in Fig. 2, and each was so rotated that the mating surfaces of the two moved in opposite directions. Loads were applied to the specimens by a spring up to 200 kg. With this highest loading it can be calculated from Hertz's formula that a maximum pressure of approximately 100,000 psi. is obtained at the center of the contacting surfaces.

The lower disk was driven at either 213 or 426 r.p.m. The upper one rotated at 90% of this speed, 192 or 384 r.p.m. The rubbing speed is,

Table I — Data on Steels Tested

	S.A.E. 1015	Oil Hardening Die Steel	Graph-Mo
Chemical analysis			
Carbon	0.11	0.92	1.49*
Manganese	0.40	1.11	0.40
Silicon	0.17	0.14	0.86
Chromium	0.05	0.46	0.16
Nickel	0.12	0.09	0.14
Molybdenum	0.02	0.02	0.24
Vanadium		0.20	
Tungsten		0.61	
Graphite			0.69
Hardening	None	1450° F.†	1480° F.†
Tempering	None	300	300
Hardness, Brinell	90		
Rockwell		C-62.5/63.5	C-61.0

*Includes the 0.69 graphitic carbon. †Oil quenched.

Fig. 2 — Cylindrical Specimen of Die Steel (1) and Soft 1015 Steel (2) in Amsler Wear Machine

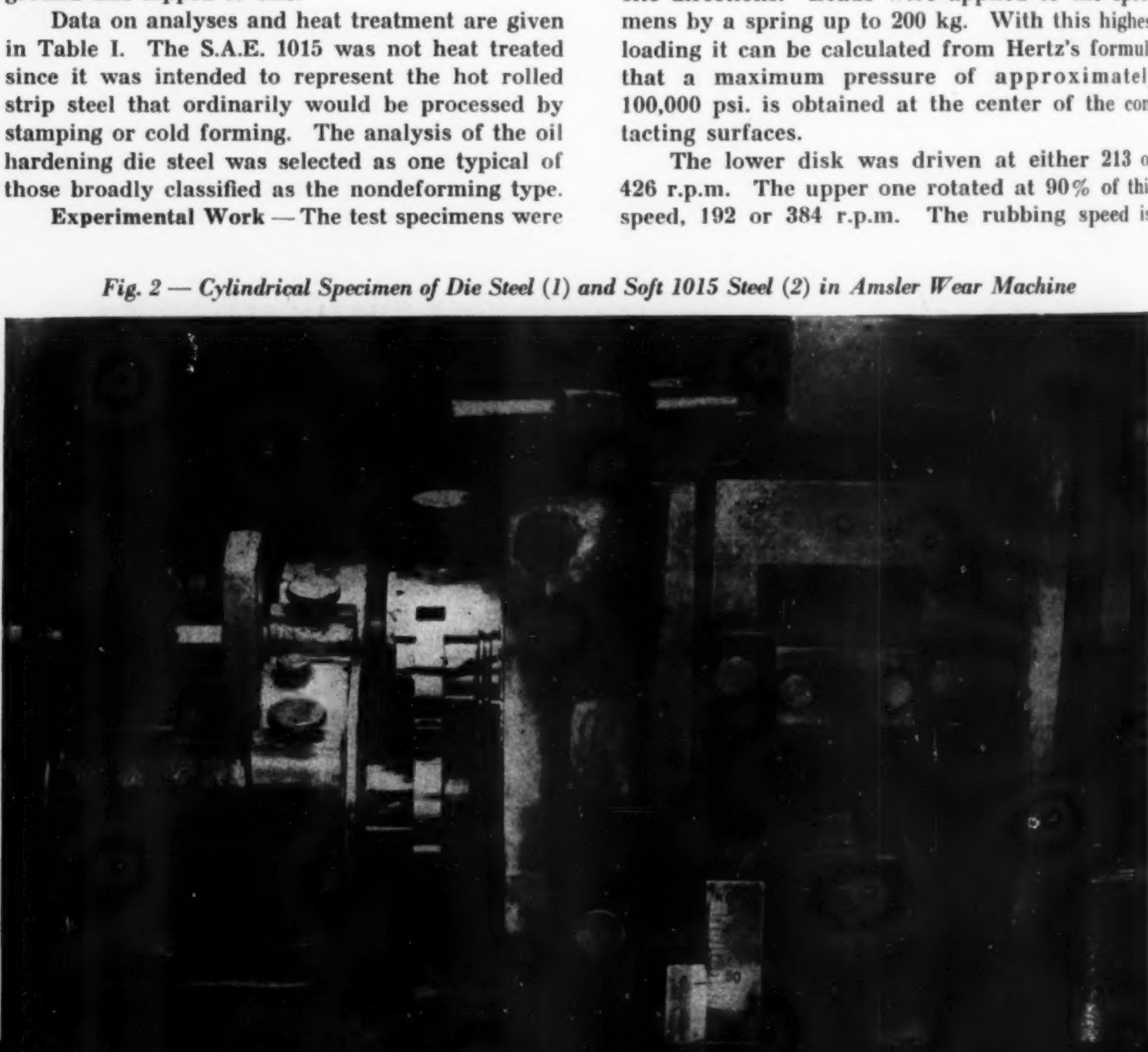


Table II — Galling Test Results

Die Steel and Test No.	Galling Load		Photographs of Specimen
	213 r.p.m.	426 r.p.m.	
Graph-Mo 1	160 kg.	—	Fig. 4
3	+200★	60-80†	Fig. 5
4	+200★	100	—
6	+200★	40-60†	—
Regular 2	140	—	—
5	120-140†	—	Fig. 6
7	80-100†	—	—
8	120	—	—

★Did not gall at 200 kg. at low speed.

†Galled while increasing from lower to upper load noted.

therefore, 212 ft. per min. at the low, and 424 at the high speed.

The characteristics of this type of test, which make it suitable for simulating the service of forming dies, are the high bearing pressures at the contact between the curved surfaces, and the comparatively low sliding velocity.

Testing Technique

The ground disks were assembled in the machine and lapped, as the first step in the test. This step was necessary to obtain an even distribution of load and uniform results. The lapping was done in successive steps with 600 carborundum, levigated alundum, and rouge, to obtain the best possible finish in the shortest time. Even so, the preparation of the specimens may take as long as 8 hr. The abrasives were suspended in a diluted extreme-pressure lubricant, to prevent galling at this stage.

As finished, both surfaces had a very uniform appearance. The hard die steels acquired a mirror finish, the soft S.A.E. 1015 steel became smooth but with a matte finish.

At the conclusion of the lapping, the surfaces were cleaned thoroughly to remove all abrasive without removing them from the machine. A pan of S.A.E. 10 motor oil was then arranged so that the bottom of the lower specimen dipped in it, and could drag enough oil up to the point of contact between the specimens to provide reproducible lubrication.

The machine was then run at the low speed, and load applied in increments. Each load was maintained for 750 revolutions of the shaft before progressing to the next load. The scale of loads was 2½, 5, 10, 15, 20, 25, 30, 40, 50, 60, 80, 100, 120, 140, 160, 180, 200 kg. The test was stopped as soon as galling occurred. If no galling occurred at 200-kg. load at the low speed, the load was removed, the speed increased and the same load sequence applied on the high speed.

Results of Tests — The galling test results are given in Table II. It is apparent from the results that the standard die steel galled at a lower load than the Graph-Mo in every run. One Graph-Mo specimen failed at a 160-kg. load at the low speed, but all of the others carried 200 kg. at this speed, and 40 to 100 kg. at the higher speed. The first test in which failure of Graph-Mo occurred at the low speed appeared to be an accidental result. Examination of the photographs of the tested specimens appeared to confirm this result, because the failure was extremely localized, being confined to a very narrow band (Fig. 4). In all other runs the zone of failure was a much wider band of badly torn material.

The standard oil hardening die steel specimens all failed at the low testing speed, at loads ranging from 80 to 140 kg.

These results indicate that Graph-Mo, in this test, will carry about twice as much load without galling as the die steel.

All of the specimens that were tested were photographed and those typical of the type of seizure obtained

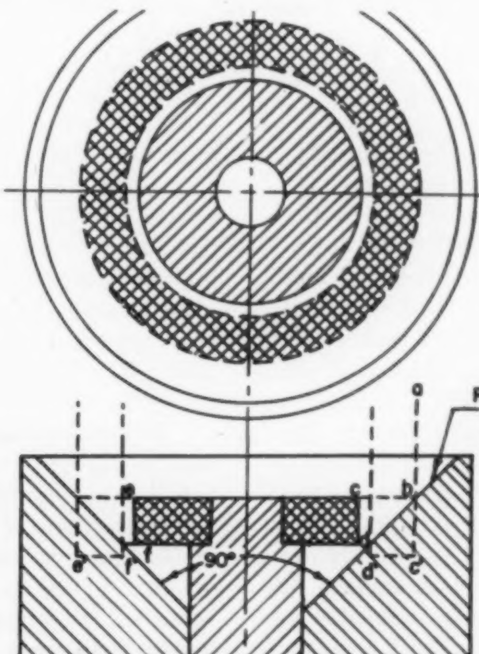
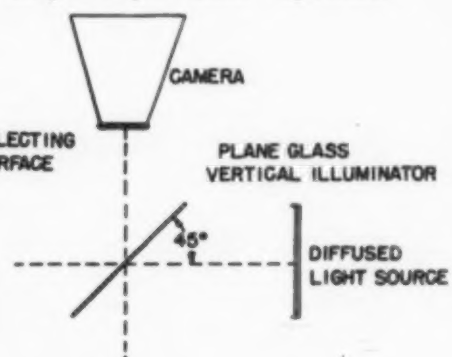


Fig. 3 — Sketch of Conical Mirror Used to Photograph the Cylindrical Surface of Amsler Specimen



are shown as Fig. 4, 5 and 6. Figure 4 was included since it shows the extremely narrow band of seizure that occurred on the only Graph-Mo sample that galled at low speed. Figures 5 and 6 are representative of the type of galling that occurred on the Graph-Mo and regular type die steels, respectively.

The photographic technique for these views is rather unusual, and requires description.

It is generally difficult to photograph cylindrical surfaces because of uneven illumination. The technique used for making the photographs reproduced in Fig. 4 to 6 avoids this difficulty by photographing a flat virtual image of the cylindrical surface.

The apparatus is sketched in Fig. 3. It consists of a 90° conical mirror, with the specimen to be photographed placed inside the concavity. Axes of cone and cylinder are coincident.

In the lower left-hand corner of Fig. 3 the behavior of the mirror is illustrated. The cylindrical surface is represented by the lines c–d and e–f. The reflections of these lines in the conical mirror are the lines c′–d′ and e′–f′, which are in the same plane, as are the reflections of all elements of the cylindrical surface.

Viewed from above, the appearance of the system was as shown in the sketch in the upper left-hand corner of Fig. 3. The top surface of the cylindrical disk is shown crosshatched. The doubly crosshatched annulus which surrounds the disk is the reflection of the cylindrical surface of the disk. The complete photographic apparatus is shown in the sketch on the right of Fig. 3.

Examination of the photographs showed that the zone of failure is similar for all tests except the first one. In most of them, the zone of failure covered a fairly wide portion of the tested surface, and consisted of torn and dragged material. In the first test, the failed area was a very narrow zone, leading to the belief that the failure was accidental and not truly representative of the material.

Fig. 4 — Galling in Narrow Band on Graph-Mo

Fig. 5 — Typical Galling in Wide Band on Graph-Mo

Fig. 6 — Typical Galling on Regular Die Steel

Microscopy and Hardness

Microstructures of the specimens of the die steel and the 1015 disk, taken normal to the contacting surfaces after completion of the tests, are shown in Fig. 7 and 8. These samples were chromium plated to preserve the original surface intact; the plating appears as a white band at the top. Those reproduced are typical of the condition of each group. Graphite in the Graph-Mo steel was quite apparent in the microstructure of the test disk, indistinguishable in structure from that shown in Fig. 1 (p. 65). The low-carbon steel picked up on the surface of the regular oil hardening die steel is shown clearly in Fig. 7. (This same condition also prevailed on the Graph-Mo steel but required greater loading to produce, as indicated in Table II.) The severely cold worked condition at the surface of the S.A.E. 1015 is depicted in Fig. 8.

Vicker's hardness readings of the S.A.E. 1015 steel disks taken at the surface and at increments below it gave values which, by conversion, were 148 Brinell on the surface, decreasing to 99 Brinell at the quarter-section. Comparing this with the original hardness of 90 Brinell noted in Table I indicates that the severe loading caused some hardness increase throughout the entire test specimen. The hardened Graph-Mo and regular die steel disks gave hardness readings practically the same as those obtained prior to testing.

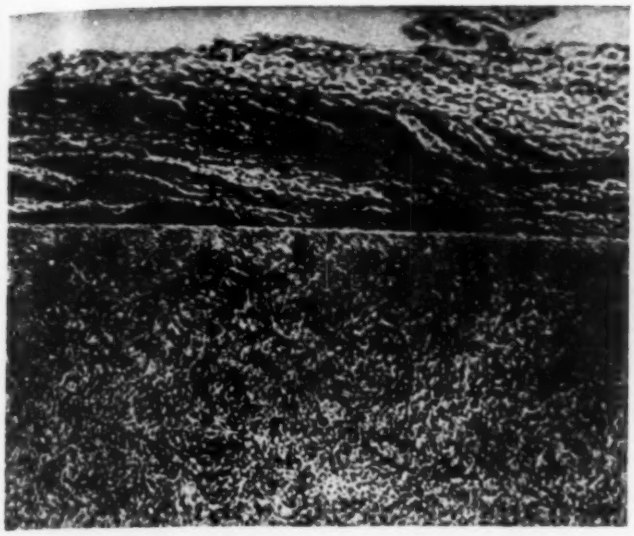


Fig. 7 — Regular Die Steel (Tempered Martensite and Carbide) at Bottom; Badly Distorted Steel Pick-Up on Surface; Chromium Plate at Extreme Top. 750×

Fig. 8 — Radial Section of S.A.E. 1015 Disk After Galling Test. Granular pearlite and ferrite, heavily cold worked at surface. 750×

Conclusions

In conclusion, these tests have shown that under test conditions simulating, insofar as possible, cold forming die operations, Graph-Mo required approximately twice the load to gall against soft steel as a conventional type of die steel. Since the hardness and other structural conditions of the materials were for all practical purposes the same, it is believed that this difference in galling resistance is attributable to the presence of graphite. The exact mechanism by which this improvement takes place is not understood, but may be associated with any or all of the following hypotheses:

1. The graphite particles serve as a lubricant to mitigate initial scuffing.
2. The graphite pockets act as tiny reservoirs to retain the externally applied lubricant even under high pressures.
3. The graphite pockets can hold any minute dislodged particles or other contamination externally introduced.
4. The hard matrix or asperities interspersed between graphite pockets serve as minute hard bearing areas that reduce friction.

These quantitative laboratory test results corroborate numerous shop production tests made on Graph-Mo dies, and are believed indicative of the greater galling resistance of graphitic steel in service as forming dies.

Correspondence

Industrial Metals of High Purity

PARIS, FRANCE

To the Readers of METAL PROGRESS:

For a long time high-purity metals were considered as laboratory products, obtained or prepared in small quantity for scientific ends. Today several laboratory processes have been built up to the industrial scale so that metals are now produced by the ton with a degree of purity formerly obtained only in the laboratory.

The methods employed almost all involve either the gaseous or ionized states: Sublimation or distillation for calcium, magnesium and zinc, carbonyl gas for iron, aqueous electrolysis for copper, fused salt electrolysis for aluminum. The purity obtained exceeds 99.9% and sometimes exceeds 99.99%.

At this degree of purity, chemical properties (resistance to corrosion) or mechanical properties (such as ductility) are obtained which are remarkable, and one is tempted to regard average industrial metals as alloys when compared with the extra pure metals.

These pure metals are indispensable in evaluating the importance of different impurities in industrial metals. Serious errors have been made in setting specification limits based on inexact knowledge of the effects of impurities.

Small amounts of impurity in a metal are not necessarily detrimental; they can be useful. For example, of the two principal impurities in industrial aluminum, one, iron, interferes with the age hardening of certain alloys, while the other, silicon, is a hardening element when combined with magnesium. Only methodical study, taking pure metal as a point of departure, can inform us about specific effects such as these.

The importance of this study of pure metals is not a new idea. More than 50 years ago Osmond undertook the study of pure iron and of monocrystals, believing this indispensable to the understanding and explanation of the characteristics of industrial products. Here, again, he was the predecessor of moderns who followed his path, but with incomparably better equipment.

The variation of a property in terms of percentage of impurities is sometimes linear, but often accelerated — for example, the magnetic permeability of iron in terms of the percentage of oxygen or carbon. It can also be a discontinuous variation as in the corrosion of aluminum by hydrochloric acid.

Extrapolation toward zero impurity is a delicate operation, and it is sometimes preferable to identify the purity of a metal in terms of the value of a property that is known to be sensitive to impurities. For example, the thermoelectric force of platinum is used as a criterion of purity.

What is commonly called deoxidation of a metal includes two very different operations: one, the reduction of the total percentage of oxygen; the other, the placing of oxygen in a less detrimental form. To do this, reducers — improperly called deoxidants — are used. A well-known example is the action of manganese and of magnesium on nickel that contains sulphur as an impurity.

The introduction, as a condition of industrial acceptability, of a specification limiting the determined impurities is therefore an intricate task, in no sense a simple administrative formality. Exaggeration in this matter only paralyzes production and increases the cost unnecessarily.

ALBERT PORTEVIN

Selective Annealing of Copper Alloys

ZURICH, SWITZERLAND

To the Readers of METAL PROGRESS:

The obviously satisfactory results obtained with the selective annealing process recommended by L. H. Seabright, in the November 1947 issue of *Metal Progress*, are certainly limited to a small number of parts. During my experience in the brass industry we found that only in exceptional circumstances could the required hardness be obtained by such an annealing procedure.

Annealing curves for copper alloys show an abrupt decrease in hardness as soon as recrystallization begins. For this reason it is difficult to obtain a certain hardness number in the range of half-hard temper. Softening depends not only on the temperature but also on the duration of heating, small variations in composition, and the previous rolling and annealing procedures. In this connection the reader may refer to two articles by M. Cook in the *Journal* of the Institute of Metals (V. 70, 1944, p. 159; and V. 73, 1946, p. 1).

Mr. Seabright mentions that 1 hr. per in. of load was allowed for heat penetration. This value varies greatly with the form of the metal that is being annealed, and therefore the value given must be considered as restricted to a small number of articles. Furthermore, the rate of flow of heat in the metal is difficult to control. Even in simple parts the metal may be hot on one side while it is considerably cooler at another place.

After selective annealing, the metal will contain both recrystallized and unrecrystallized grains. This structure will never be beneficial.

Mr. Seabright's practice can by no means be recommended.

O. H. C. MESSNER
Consulting Engineer

Mr. Seabright Replies

It seems that Dr. Messner has approached the problem of annealing from the standpoint of mill operation, whereas I have outlined the procedure of a manufacturer that is forced to use material available in warehouse stocks. If a system such as I have outlined is not used, a manufacturer may find himself without material and faced with costly delays in production until such time as stock of the desired temper is available.

I agree that there are certain grades of brass and other nonferrous alloys in which softening is quite rapid as recrystallization begins, and with such material it is rather difficult to anneal within the limits of the half-hard temper. However, for

yellow brass, nickel-silver spring stock, and 5% phosphor bronze, the hardness decreases gradually with annealing temperature, as shown in the diagrams on pages 71, 202 and 268, respectively, in the book "Copper and Copper-Base Alloys" by Wilkins and Bunn.

We have found that after a few preliminary tests it is possible to establish an optimum cycle which, with temperature control of plus-or-minus 5° F., will give the desired hardness within plus-or-minus two points on the Rockwell B scale, for the alloys mentioned in the preceding paragraph. The material is purchased in accordance with A.S.T.M. specifications, so that each batch will be uniform in composition, temper and grain size.

Subzero Treatment of Bearing Parts

South Bend, Ind.

To the Readers of Metal Progress:

In the May issue of *Metal Progress* there was an article, "Subzero Treatments for Carburized Medium-Alloy Parts" by J. C. Selby and E. S. Rowland. The data in the article were obtained from samples of nine grades of steel that had been carburized, quenched in oil, cooled to subzero temperatures (both with and without intermediate tempering), and then tempered. Three of the steels, Krupp, 4620 and 4320, are regularly used for roller bearings. One conclusion in the article was that, because of expense, subzero treatment does not seem to offer any advantage in the processing of carburized and ground products, such as roller bearing parts.

It is improbable that carburized bearing parts are being made with only a single-quench heat treatment such as was applied to the samples for the article. As most carburized bearing parts are double quenched, there are fundamentals not covered in the article that should have consideration.

Before a part is quenched at the end of a carburizing cycle, all the carbon in the case is in solution in the austenite. At times, Krupp and 4320 steels may have some excess carbide in the zone adjacent to the surface, but this does not affect the ultimate part as the outer zone is removed by grinding. Because of the high carbon content of the austenite — usually 0.85 to 1.00% C — the M_s point is close to 200° F. and M_f is far below the temperatures reached with commercial subzero apparatus. Thus, with a single quench, considerable ausfenite is retained even after the steel has been cooled to −100° F. It is not surprising, therefore, that the authors failed to transform enough austenite to get hardness values approaching C-65.

By using a double-quench practice combined with a spheroidizing treatment between quenches, it is possible to control the amount of carbon dissolved in the austenite prior to the second quench. The result is that M_s is raised to about 400° F. and M_f to about −65° F. The undissolved carbide provides additional wear resistance. By raising M_f, full hardness can be obtained with commercial subzero apparatus.

Because the bearing steels, especially the Krupp grade, may contain retained austenite even with rigid control, any treatment that will increase the extent of the austenite-to-martensite transformation without affecting hardness is desirable. Tempering treatments for this purpose affect the hardness, and there is some question whether enough austenite is transformed by tempering. As martensite can be formed by surface deformation, there is danger of its formation from any retained austenite by the shock loading to which many bearing parts are subjected. Some bearing manufacturers advocate heating the parts in oil after a brief service period, in order to temper such martensite as may have been formed. This procedure is not desirable from the standpoint of size stabilization.

This discussion is intended to correct any interpretation from the article on subzero treatments that bearing parts have relatively simple heat treatments. The expense involved in subzero treatment is a separate issue. With the constant demand for bearing parts to possess increased hardness combined with increased shock resistance and size stability, the advantages of subzero treatments, when M_f is controlled at a reasonable level, cannot be overlooked.

H. Habart
Chief Metallurgist
The Torrington Co.

Dr. Rowland Replies

Mr. Habart is correct in stating that most carburized bearing parts are double quenched. The introductory paragraphs of our article were intended to emphasize that our data are applicable only to single-quench techniques, which are used when the distortion must be held to a minimum.

We are still firmly convinced that the factors (excepting stress) controlling the retained austenite in the hardened case of double-treated parts are the carbon distribution in the case as a result of the carburizing practice, the carbide size and distribution during the hardening treatment, the austenitizing conditions, and the quench. If these factors are adequately controlled, corrective measures such as subzero treatment are unnecessary.

Alloy Steel has assumed such a tremendous role of such primary importance in so many essential industries, that the American Society for Metals welcomes this opportunity to dedicate as the theme of the 30th National Metal Congress & Exposition — "A Salute to Alloy Steel".

Alloy Steel is a truly American achievement. Since the tremendous improvement in Alloy Steel during the past 30 years has paralleled the growth and importance of the American Society for Metals, it is fitting that the 20,000 members of this society, who represent the Alloy Steel producing and consuming industries, should sponsor this well-deserved tribute.

BRILLIANT TECHNICAL PROGRAM

A vast panorama of progress in Alloy Steel production and fabrication will be presented at the Philadelphia Metal Congress & Exposition. A brilliant program of over 100 stimulating and helpful technical papers on all aspects of Alloy Steel will be presented by recognized authorities at the Metal Congress.

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Along with the distinguished displays of over 350 manufacturers and processors of metals and metal products will be a spectacular dramatization of the 75 years progress in Alloy Steel. Occupying the entire stage of Convention Hall, this exhibit will give visitors to the Exposition the complete story of Alloy Steel.

From all over the nation, metal men are making plans to come to Philadelphia for this great jubilee. Don't miss it—make your arrangements TODAY!

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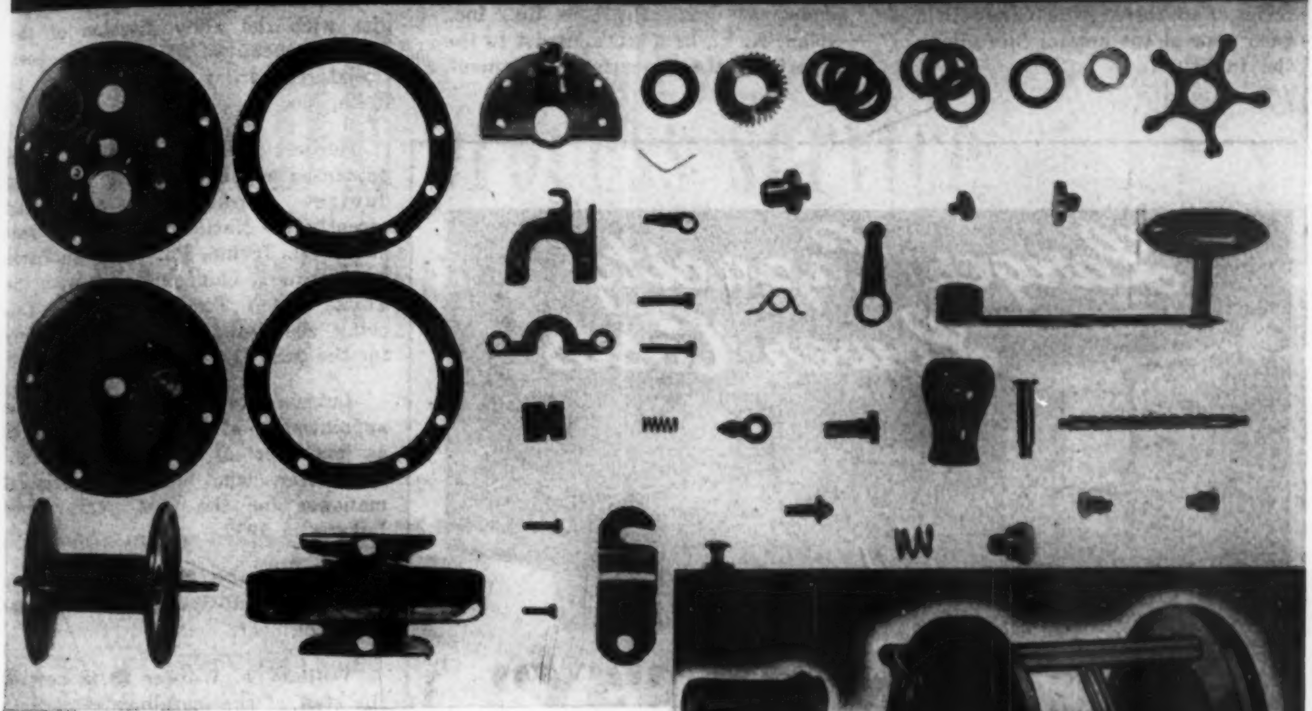

Parts for a Penn Reel. Many of these are standardized and are interchangeable regardless of type or size of reel.

GREETINGS to all fishermen from Revere and the Penn Fishing Tackle Mfg. Co.! Revere is glad you like Penn Reels, for though you may not realize it, they contain several different Revere Metals, each one chosen to match specific requirements. A reel is a complex machine, and it has to be well designed and accurately made of fine materials if it is to serve you well. Penn specializes in salt water reels, and during the past three years alone has bought the following from Revere:

Brass Strip, 11 different items, for 12 parts
Brass Rod, 14 different items, for 13 parts
Phosphor Bronze Strip, 6 different items, for one part
Brass Tube, one item for one part

Brass is specified for its easy workability, corrosion resistance, and excellent plating qualities; Phosphor Bronze for its wear resistance as used in the drag ratchet dog in several sizes of reels. Complete parts for a reel are shown above; some of them, making use of Revere Metals include: handle, level wind worm, pinion yoke, gear plate with drag ratchet, drag adjusting star, spool bearing, front and back plate rings . . . Revere is always delighted when its products contribute to the enjoyment of sport; it gives us a sense of added accomplishment, because in most cases the Revere Metals have utilitarian end uses. But whatever it is you make, whether for sport or not, remember that it is our business to serve you with fine metals, and to collaborate with you in your search for economy in the specification and use of those metals.

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Personals

The Institute of British Foundrymen has invited **Harry A. Schwartz**, director of research for National Malleable and Steel Castings Co., Cleveland, to address the organization in London on June 9. His lecture, "Solved and Unsolved Problems in the Metallurgy of Blackheart Malleable", was the tenth in the Edward Williams series of addresses, which are delivered each year at the annual conference of the Institute.

William H. Nikola has joined the Westinghouse Electric Corp., at Jersey City, N. J., as quality control engineer.

On receiving his degree in metallurgical engineering from the Colorado School of Mines, **David W. Reese** is now with Ingersoll-Rand Co.

E. Leroy Aul, formerly research assistant at Case Institute of Technology in its research laboratory for mechanical metallurgy, is now employed at Clark Brothers Co., Inc., Olean, N. Y., as a metallurgist in the experimental engineering department.

D. P. Antia is now development officer for metal industries with the Ministry of Industry & Supply, India. He has also been appointed honorary secretary to the Indian Institute of Metals.

Following graduation from the University of Notre Dame, **J. C. Kremer** has joined the research laboratories division of General Motors Corp., Detroit, as a physicist.

Thurston D. Brown, metallurgist with the alloy division of the U. S. Bureau of Mines, has been transferred from Salt Lake City to Rolla, Mo.

Herbert L. MacBride has been appointed manager of the precision devices department established recently at National Forge & Ordnance Co., Irvine, Pa. Mr. MacBride was formerly chief engineer with the Tinius Olsen Testing Machine Co., being associated with that company for the past 11 years.

Lukens Steel Co. announces the appointment of **L. P. McAllister** as manager of steel plants. Mr. McAllister, who has been assistant manager for the past year, joined Lukens in 1922.

W. N. Smith is now affiliated with Universal-Cyclops Steel Co. as sales engineer.

William T. Warner is now on the staff of the metallurgical section, nuclear reactor project, of the Brookhaven National Laboratory, Upton, N. Y.

Wilton F. Melhorn is now assistant professor of metallurgy at Virginia Polytechnic Institute, located in Blacksburg, Va.

Douglas W. Grobecker has joined the staff of the chemistry and metallurgy division of Los Alamos, N. M., Scientific Laboratory of University of California. He was formerly with American Brake Shoe Co., "Amsco" division.

Universal-Cyclops Steel Corp. has appointed **Lorenz W. Rinek** as district sales manager of the company in Indianapolis, Ind.

Federated Metals announces the appointment of **Raymond A. Quadt** as assistant manager of its general aluminum department. He will continue supervising aluminum research at the central research laboratory of American Smelting and Refining Co., the position he held before his promotion.

Personals

Frederick Stirbl, formerly research metallurgist for Crucible Steel Co. of America, has established a metallurgical consulting practice in New York City.

A. J. Stimac is now machinist foreman of the Denver Tramway Corp., Denver, Colo.

C. H. Betts has joined the staff of the Physical Metallurgy Research Laboratories of the Bureau of Mines, Ottawa, Canada.

Stuart R. Vandenberg has accepted a position in the general engineering and consulting laboratory of the General Electric Co., Schenectady. Prior to accepting this position, Mr. Vandenberg has been connected with Westinghouse Electric Corp. and more recently has had special assignments for General Electric.

Temporarily giving up his consulting engineering practice in South Dakota and Nebraska, **Gale S. Hanks** has joined the staff of the Kennecott Copper Corp. at its reduction plant in McGill, Nev., as assistant gas chemist of the copper smelter.

After serving in the conservation division of the War Production Board, in the Navy Department and at Chefford Master Mfg. Co. for the past six years, **J. Ralph Fritze** has returned to Hotpoint, Inc., Chicago, where he will be engaged in development work.

Following 20 years of association with American Cyanamid Co., **George D. Johnston** has resigned and is now with Commonwealth Industries, Detroit, as plant superintendent.

E. Wayne Everhart has resigned his position with the Glenn L. Martin Co. to head the corrosion laboratory of Permanente Metals Corp.'s newly formed research department at Spokane, Wash.

S. S. Gill, who has been in the United States for the past two years studying the steel wire industry on assignment for the government of India, is returning to his homeland, where he will work with the Indian Steel & Wire Products, Tata Nagar, India, as metallurgist.

Formerly stainless and toolsteel metallurgist in the mill metallurgical department of Timken Roller Bearing Co., **R. E. Groethe** is now metallurgical engineer with Corning Glass Works, Corning, N. Y.

Personals

J. L. McCloud has been appointed director of chemical engineering and chemical and metallurgical research at Ford Motor Co., Dearborn, Mich., to succeed the late R. H. McCarroll. Mr. McCloud, who has been with Ford since 1915, was previously assistant director.

Permanente Metals Corp., Spokane, Wash., announces the appointment of **R. E. Peterson**, formerly of the Fansteel Metallurgical Corp., as director of the X-ray diffraction laboratory of Permanente's new Kaiser aluminum metallurgical research department.

James L. Griswold, formerly chemist at Bethlehem Pacific Coast Steel Co., is now metallurgist at Pittsburg, Calif., works of Columbia Steel Co.

W. C. Motz has joined the Babcock & Wilcox research laboratories at Alliance, Ohio, in the capacity of assistant metallurgist.

Jack P. Morrissey has been transferred from the Fort Wayne works laboratory of General Electric Co. to the technical laboratory of the Bloomfield, N. J., works. He is an engineer in charge of welding and the development of welding processes.

After receiving his M.S. from the University of Illinois, **Charles R. Cook, Jr.,** has accepted a position as research metallurgist with the titanium division of the National Lead Co., South Amboy, N. J.

A. M. Castle Co. has transferred **Geo. S. Brett** from sales manager at Seattle to manager of the Castle plant in Kansas City, Mo.

The Westinghouse Electric Corp. has transferred **Ernest L. Layland** to its Sturtevant division, Hyde Park, Mass., as materials and standards engineer.

George Bennett, who was plant development engineer for M & M, Ltd., Newark, N. J., is now associated with the Harold F. Howard Co., Detroit, as consulting engineer. Before joining M & M, Mr. Bennett was chief tool engineer for Fruehauf Trailer Co.

Following the completion of work for his Sc.D. degree from Massachusetts Institute of Technology, **R. J. Teitel** has taken a position as engineer on the staff at Brookhaven National Laboratories, Upton, N. Y.

Personals

Allegheny Ludlum Steel Corp. announces the transfer of **H. N. Arbuthnot**, assistant to the president, from Detroit to Washington, D. C., where he will serve as representative of the company in relationships with the various governmental agencies and sales manager in the Washington area. Mr. Arbuthnot joined Allegheny Steel in 1926 and has served in various sales capacities since.

George F. Comstock, for many years chief metallurgist of the Titanium Alloy Mfg. Co., Niagara Falls, N. Y., has been appointed assistant director of research for the Titanium Co.'s laboratories.

Richard Doughton, Jr., has been appointed assistant to the general manager of the Hallett Mfg. Co., Inglewood, Calif.

C. I. Hayes, Inc., announces the appointment of **Leonard J. Edwards** as New England sales representative, with offices in Providence, R. I. Mr. Edwards was for many years field sales manager for General Alloys Co.

Russell C. Fancher has been appointed sales representative in the Detroit area for U. S. Stoneware Co. of Akron, Ohio. He has been associated with several chemical and metal-finishing companies in Detroit, most recently with the Eaton Chemical & Dyestuff Co.

E. C. Troy, formerly vice-president of research and development of the Dodge Steel Co., will represent the Foundry Equipment Co., International Molding Machine Co. and National Engineering Co. as eastern sales and service engineer with offices in Palmyra, N. J.

Ohio State University's Lamme Medal for meritorious achievement in engineering was awarded this year to **Earle C. Smith**, chief metallurgist of Republic Steel Corp., Cleveland.

V. W. Wiskochil has resigned his position at Revere Copper & Brass, Inc., to become metallurgist with the Illinois Watch Case Co., Elgin, Ill.

Leonard W. Walline, formerly observer at the Carnegie-Illinois Steel Corp., is now metallurgist at the Motor Wheel Corp., Lansing, Mich.

Michigan Steel Casting Co. announces the appointment of **Fred H. Currie** and **S. G. Higginbotham** as sales representatives in the states of California and Arizona. Their headquarters will be in Los Angeles.

Herbert Macmillan has recently founded the Herbert Macmillan Co. of New Rochelle, N. Y., to manufacture hardware and other yacht products, and to distribute Korodless stainless steel ropes and fittings, and other specialties.

Tennessee Coal, Iron and Railroad Co. has promoted **R. W. Mueller**, who has been with the company for the past 14 years in various sales capacities, to the position of manager of sales of the stainless steel division.

Carnegie-Illinois Steel Corp. announces the appointment of **J. Douglas Darby** as vice-president in charge of sales. Mr. Darby was connected with Alan Wood Steel Co. for 20 years and joined the sales department of Carnegie-Illinois in 1939.

Gordon L. Meeter, formerly metallurgist for Solar Aircraft's Des Moines plant, has joined the Babcock & Wilcox Co.'s research and development department at Alliance, Ohio, as metallurgist.

Vernon R. Scott is now metallurgist for Dickson Weatherproof Nail Co., in charge of metallurgical problems for the main plant in Evanston, Ill., as well as branch plants in Birmingham, Ala., Houston and Galveston, Tex., and Los Angeles.

Adam Baer is now secretary-treasurer of Global Electric Mfg. Co., Inc., Stockton, Calif.

Robert T. Burnham, formerly with the Buick Motor Division, General Motors Corp., is now at Columbia University working toward his M.S. degree.

N. W. Richardson, formerly chief coordinating engineer for Colorado Fuel & Iron Corp., is now with the National Tube Co., Lorain, Ohio, in conjunction with the operation of the newly developed mills for producing seamless tube.

Formerly chief metallurgist and manager of metallurgy and sales for Barium Steel and Forge, Inc., **James C. Hartley** has recently been appointed vice-president, general manager and director of the company.

Russell J. Nadherny, formerly chief engineer, has been promoted to the position of executive vice-president of Barnes and Reinecke, Inc., Chicago.


First Cost isn't
everything

Personals

E. C. Jasper, formerly with Central By-Products Co., Mapleton, Minn., is now metallurgist for the Victor Adding Machine Co., Chicago.

Hugh D. Luke has joined McKinsey & Co., management consultants of Boston, as consultant in manufacturing. He was previously manager of the Mount Hope Machinery Co., Taunton, Mass.

Claud S. Gordon Co. has appointed **Lloyd J. Bohan** as its representative in the Los Angeles area. Mr. Bohan is a former secretary of the Chicago Chapter of the American Society for Metals.

Ohio Ferro-Alloys Corp., Canton, Ohio, announces that **R. L. Cunningham** has been elected president of the company, the position vacated by the death of **L. G. Pritz**. Mr. Cunningham was previously executive vice-president. His former position will be taken by **W. W. Pritz**, who had served as works manager and vice-president.

Allegheny Ludlum Steel Corp. has appointed **Truman B. Brown**, former assistant manager of cutting-steel and toolsteel sales in Pittsburgh, as assistant district manager of sales for the Detroit territory. **Roger S. Ahlbrandt**, formerly district manager of sales in Pittsburgh, will succeed Mr. Brown as assistant manager.

George Sachs, director of the research laboratory for mechanical metallurgy and professor of physical metallurgy at Case Institute of Technology, Cleveland, has been appointed director of the National Metallurgical Laboratory of India located at Jamshedpur. He will assume his new position on Oct. 1, 1948. He is being succeeded at Case by **William Marsh Baldwin, Jr.**, who will be given the title of research professor of metallurgy. Dr. Baldwin has been associated with the Chase Brass & Copper Co. in Cleveland.

Adam J. Texter has resigned from the Exothermic Alloys Sales and Service Corp. to accept a position with the Firth Sterling Steel and Carbide Corp., McKeesport, Pa., where he will be in charge of melting.

Clyde J. Hibler, formerly with Surface Combustion Corp., Toledo, Ohio, has established a consulting service for the preparation of technical catalogs, manuals and publicity, in Chicago.

Brittle Fracture in Mild Steel Plates

A CONFERENCE of engineers and metallurgists was held in October 1945 at the Engineering Laboratory of the University at Cambridge, England, attended by 51 persons representing eight British institutions and governmental agencies, a number of private firms, and the U. S. Coast Guard. The discussions were limited strictly to the problem of brittle fractures, which occurred quite frequently in ships built during the war. These used mostly American-made steel and American welding methods. Five of the papers presented were written by metallurgists, one by an engineer-builder, and one by a member of the U. S. Coast Guard. The following represents a series of abstracts of those papers, as they appeared in *Engineering* in December of 1947 and later.

Brittle Ships

Prof. J. F. Baker,* in a discussion of "The Problem of Brittle Fracture in Ship Structures", noted that stresses in ship structures escape precise analysis. The shipbuilders of the past proceeded from certain reasonable assumptions used and appropriate safety factors. Failures in riveted ship constructions were rare, in spite of considerable residual stresses; such fractures as did occur were of the ductile type and localized. Computation of welded structures proceeded along the old lines, but large fractures were frequent and mostly of the sudden, brittle type.

Residual stresses may form a contributing factor, but their presence is of secondary importance. [It appears that Professor Baker would rather eliminate them entirely from consideration.] In his opinion, static systems of stresses applied to ductile steel never result in brittle fractures. A plastic deformation always precedes failure. A triaxial system of stresses of fairly equal intensity might prevent the development of sufficient shear stresses and bring about a brittle fracture, but this point cannot be tested experimentally and the existence of such a system of stresses over large areas is most unlikely. *(Cont. on p. 86)*

**Engineering*, Dec. 5, 1947, p. 532.

Brittle Ships

(Cont. from p. 84) Such a system may form, locally, a contributing factor, but it is not the main factor.

Likewise, the biaxial stressing of the ship plates in their own planes cannot account for the observed absence of plastic deformation in the fractures. Neither can any theory of the notch effect, advanced so far, explain the brittle behavior in mild steel without contradicting a number of experimental facts. An apparently satisfactory plate passing the normal tensile test, may show low capacity for local plastic deformation under an analogous test in the presence of a notch, the depth of which amounts to about 4% of the width of the piece tested. However, the criteria permitting one to divide steel plate into the brittle and the ductile classes have not yet been found.

While brittle fractures in ship plates have nothing in common with true fatigue fractures, the presence of fatigue under the action of temperature variations and of the battering by waves might form a serious contributing factor. Resonance effects, due to vibrations of machinery, might have a greater effect in the continuous welded hull than in the old riveted structure and accentuate the residual stresses. Experiments indicate, however, that superimposed alternating stresses are likely to *reduce* residual stresses, not to increase them.

It has been found that, if a welded ship structure has enough ductility, it resists the action of nearby underwater explosions much better than the riveted structure does.

Low Ductility in Testing*

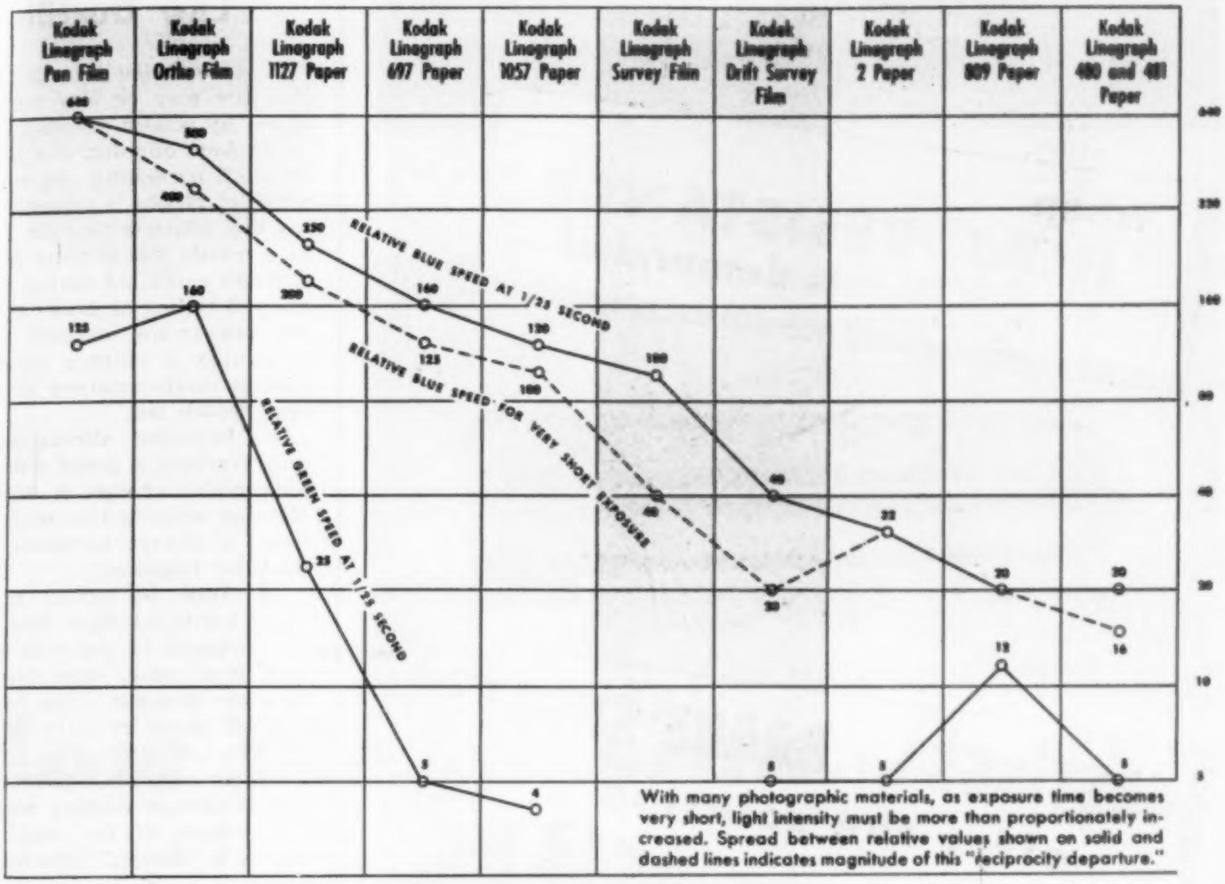
Professor Baker also discussed the "Causes of Low Ductility in Mild Steel". This article boils down to the presentation of the things *we do not know* regarding the causes of failure, to wit:

1. Brittleness is indicated by low elongation, low reduction of area, and low energy absorption in the notch-bar impact test.

2. Very little is known about the conditions leading to fracture.

3. Fracture does not always occur on the glide planes, as theoretically predicted. *(Cont. on p. 88)*

**Engineering*, Dec. 5, 1947, p. 548.



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Low Ductility

(Continued from p. 86)

Fracture may or may not be preceded by plastic deformation.

4. Low ductility can be caused by work hardening. An instructive series of graphs is presented, showing that ultimate strength exhibited in a tensile test is away below the strength exhibited during the processes of rolling or drawing, because the maximum amount of work hardening of which a metal is capable is rarely attained in an ordinary tensile test.

5. Repeated alternating loads may fracture a metal without any appreciable change in dimensions. Material subjected to such alternations is always hardened and its ductility impaired.

6. Tests on metal previously work hardened show that increasing amounts of previous deformation first raise, then lower the ultimate strength. The interpretation of these results is still in dispute.

7. In notch-bar tests the fractured surface leading away from the bottom of the notch always shows a "fibrous" structure, while the balance of the area is crystalline. In slow bend tests the fracture may contain zones of fibrous structure alternating with those of crystalline structure. The first correspond to the rapid, the second to the slow drops of the load in the final stages of the deformation.

8. Constitutional and structural changes due to cooling from high temperature may produce brittleness either by chilling or by strain-aging.

9. Combination of notch effect with low surrounding temperature might cause an enormous loss of ductility. Ordinary impact tensile tests show a rapid drop of ductility in ordinary mild steel below −150° F. Notched impact test shows the same between +70° and zero F.

10. High-velocity impact test may double the yield point without affecting tensile strength and elongation even in the absence of a notch. If the piece is notched, the notch can change a fibrous structure into crystalline.

11. Raising the test temperature has little influence upon cleavage stress and a great influence upon the shear stress. At some temperature the two are equal. Above it the metal fails plastically by shear; below it the fracture is brittle. *(Cont. on p. 90)*

Low Ductility

(Abstract starts on p. 86)

12. Under the same specific loading and all other identical conditions a narrow specimen undergoes a greater deflection and fails with a fibrous fracture, while a large specimen breaks suddenly and shows a crystalline fracture.

13. Stress-corrosion and fatigue-corrosion may cause the appearance of the notch effect.

To SUMMARIZE — There are two ways by which a steel may fracture. Shear is the normal way. Cleavage — associated with high velocity of propagation and low energy absorption — is the other way. The exact conditions necessary to start such fractures remain to be determined.

Brittle Ship Steel*

J. L. Adam, an engineer rather than a metallurgist, presented the general picture of failures that came to his attention from the time preceding the first and up to the end of the second world war. He mentions the fact, that while riveted ships were frequently built of badly segregated and laminated steel plates, catastrophic or dangerous fractures were practically unknown. They became frequent only with the start of welding.

He mentions eight types of fractures mainly differing one from another by the spot they start in. Only one or two could be directly traced to brittle steel. One or two types are apparently considered unavoidable, none too dangerous, and can be repaired reliably.

[The article contains an interesting diagram of the location of the types of fractures described, but the fractures themselves are described in a perfunctory manner. No attempt is made to analyze the stresses that might have acted in those locations.]

Notch Brittleness†

E. Orowan of the Cavendish Laboratory, Cambridge, brought up an interesting point when he mentioned the fact that Tetmajer described the notch effect in 1885 and that this effect *(Cont. on p. 92)*

**Engineering,* Dec. 12, 1947, p. 556.

†*Engineering,* Dec. 19, 1947, p. 581; Dec. 26, 1947, p. 605.

Notch Brittleness

(Cont. from p. 90) was *rediscovered* a number of times during the 60 years that have passed!

Referring to the previous theoretical work of Ludwik, Hencky, and of his own (in 1944), Dr. Orowan states that a certain maximum ratio exists for every metal between its instantaneous yield point and its ultimate tension stress, which specifies the nature of the fracture resulting. When the tension stress equals or is lower than the yield stress of the unstrained material, the latter is quite brittle. If the tension stress does not reach three times the value of the instantaneous yield point, the material will possess notch brittleness in a decreasing degree. If the ratio is above three, notch brittleness does not occur.

Under normal conditions the brittle strength of carbon steel is close to three times its instantaneous yield point. Therefore a moderate rise in temperature eliminates notch brittleness, while a moderate drop increases it greatly. The cold brittleness of iron and steels is due to the enormous increase in the elasticity and in the yield point, which does not take place to the same extent with other "ductile" metals.

The author suggests tentatively that the tendency for notch brittleness should be studied by determining the yield stress and brittle strength. Since brittle strength is a rather elusive quantity, it might be measured by tensile tests (as close as possible to absolute zero). The latter in his opinion should correspond to the "brittle strength", which changes but little with the temperature.

[ABSTRACTER'S REMARK: Should this suggestion prove true, an analogous statement should be applicable to work hardening. Maximum strength obtainable for a work hardened metal, and the yield point corresponding to a given degree of working, should indicate the degree of notch brittleness of a work hardened metal, which is most important since a certain amount of work hardening is frequently desirable. M. G. C.]

Dr. Orowan mentions further the theoretical ideas of Ludwik, Griffith, Davidenkov and Fridman. None of these seems to be satisfactory. Fridman's theory is considered to be an advance over the others because it *(Cont. on p. 94)*

Notch Brittleness

(Abstract starts on p. 90) implies, in a rough way, that plastic yielding is a necessary condition for a ductile fracture.

"Ductile" fracture is considered to possess a wider meaning than "shear" fracture. The first embraces the second — as, for instance, is the case of the cup-and-cone fracture in steel. The fracture of the central part is a normal fibrous fracture.

After describing an interesting experiment with a thin disk of tin welded in sandwich fashion between two bars of steel and pulled in tension (in which the soft disk underwent no plastic deformation and withstood a stress 5.5 times its ultimate, fracturing at the contact surface of the weld only), Orowan examines the ideas of Kuntze and McAdam on cohesive stress, but finds that neither helps much, while each complicates the situation. So, the phenomenon of ductile fracture remains unexplained.

In the second portion of his paper Dr. Orowan switches back to the problem of brittle fracture and to Ludwik's concept of triaxial stress at the bottom of the notch, which prevents the start of the plastic deformation before the true strength is reached. If the greatest tension at yielding is plotted as a function of the plastic strain, the curve of the notched specimen would at some depth of the latter intersect the line of strength of the material and pass beyond it.

Kuntze and McAdam extended logically the above concept for very sharp and narrow notches or cracks — which should have caused extremely high yield points and bring about a notch brittleness in *every* metal. These objections to Ludwik's idea were eliminated by Orowan when he computed that the maximum possible rise of yield point due to the notch effect is about 3.3. In most of the ductile metals this value will still be below their basic strength.

Orowan refers further to Mott's theory of the effect of the velocity of the crack propagation. But, since notch brittleness obtains both in impact and in slow bending that theory cannot hold, except (as the author states) in microscopical dimensions — meaning "within single grains".

He considers further the increases in the (Cont. on p. 96)

Notch Brittleness

(*Starts on p. 90*) yield point in the restricted volumes of a stressed plate containing numerous rivet holes. This effect, which takes place also in notches, he calls "superstressing", and states that Müller and Barbers found 10 to 20% superstress in semicircular notches and 100 to 200% superstress in sharp notches. These latter investigators used X-rays to determine the internal stresses (1935).

He considers further the plastic deformations in single grains and the Lüders lines and concludes that all phenomena leading to increased notch sensitivity (temperature, strain-aging, rate of deformation) are secondary effects; the main thing is the possibility that the yield strength may grow far beyond the true cohesive strength.

The last two questions examined by Orowan are the value of the notch impact test and the possibility of eliminating notch brittleness. He maintains that Charpy or Izod impact values may have a meaning only when comparing nearly identical materials. For ductile metals the figure is useless; for strictly brittle ones — meaningless. The main feature is the relative area of the crystalline part of the fracture. The presence of a fibrous area indicates that the metal won't fail without a preceding plastic deformation.

The second question is answered by stating that notch brittleness might be inherent and therefore could not be cured, or it might be caused by chemical or mechanical imperfections. He suggests the production of samples from spectroscopically pure components in order to answer the first part of the question, and of purposely contaminated (or defective samples containing microscopic cracks) in order to test the second part of it.

A Philosophy of Fracture*

Some theoretical considerations that arise in the study of the fracture of metals was the topic discussed by N. F. Mott in his contribution to this symposium on brittle fractures in mild steel plates. He first considered Griffith's theory

(*Continued on p. 98*)

**Engineering*, Jan. 2, 1948, p. 16.

Philosophy of Fracture

(Continued from p. 96)

of fracture in brittle solids. Its formula, $F = \left(\frac{2ET}{\pi c} \right)^{1/2}$ where F is the breaking stress, E the modulus of elasticity, T the surface tension and *c* the length of pre-existing crack, is accepted as quite valid for substances like glass, but the presence of "pre-existing cracks" in crystalline substances itself calls for additional explanation.

Professor Mott points out that the existence of natural cracks within metallic crystals is difficult to conceive [although this abstracter would here ask: "Why?"]. He develops a very simple analysis of stresses occurring in a grain that carries an incipient crack, but has to make a very large number of assumptions in order to do so. He then suggests tentatively that the crack must propagate in notch brittle, but otherwise ductile materials with a velocity near to that of the velocity of sound in that material, in order that the plastic yield could find no time to develop.

[The balance of Mott's contribution forms a philosophy of fracture, so to speak, but it produces no specific explanation applicable to the fundamental problem posed by the conference. M.G.C.]

Brittle Armor Plate*

Although armor plate can hardly be called mild steel, whose brittleness was the topic of the series of papers under review, its reaction to impacts of explosive violence may shed some light on the question at issue. D. E. J. Offord therefore presented some notes on his experience with the testing of armor plate. The metal was a Ni-Cr-Mo steel accepted under most rigid specifications. Nevertheless some plates would crack under impacts far below expectations. The reasons were never made clear. Fractures in such cases were always crystalline in nature, rather than "fibrous", as desired.

Analogous ballistic tests were made during the war with plates of mild steel, whole or joined together by butt welding. The *whole* (unwelded) plate usually bulges out and its thickness *(Cont. on p. 100)*

**Engineering*, Jan. 16, 1948, p. 53.

Brittle Armor Plate

(Cont. from p. 98) is reduced over 50%. Welded plates and plates with long beads of weld metal on their surfaces, regularly cracked with a brittle fracture. Grinding the weld down flat did not help; the plate still cracked under less than 50% of the explosive charge that would merely bulge a whole plate. Neither normalizing before welding, nor heat treating at 1200° F. afterwards was of any help.

Out of two 1-in. thick whole plates tested in the same manner, one cracked and exhibited a cleavage fracture. Thinner plates behaved better; in plates ¼ in. thick only the weld has shown a brittle fracture, while the crack in the body of the plate was of a ductile nature.

The author suggests tentatively that the thickness of the base plate and the lack of continuity in geometrical or structural conditions in the welded plates might account to some extent for brittle fractures.

Effect of Steel "Quality"*

Cracks in welded steel ship structures are not necessarily sudden, in the opinion of E. M. MacCutcheon of the U. S. Coast Guard. Instances are known where the cracks were slowly advancing under the action of waves. "Continuity" effects are quite definite. Even a partial, shallow weld between two plates can form a link for the crack to propagate across.

Of 4700 ships built by welding methods in U.S.A. during the war, 20% suffered structural failures, half of them serious. Most were reported in the January 1946 issue of the *Welding Journal.* The situation was remedied to an extent by riveting "crack arresters" to the plates in dangerous positions.

MacCutcheon further related the experiments under way at present in the U.S.A. on various heats of mild steel. He mentioned test pieces with an internal notch, tests on plate widths forming a geometric progression starting at 3 in. and ending at 72 in., tests on fabricated specimens of ¾-in. plates weighing two tons. He stated that fully killed and normalized plates were better than rimmed, either "as rolled" or as strain-aged. *(Cont. on p. 102)*

**Engineering*, Jan. 23, 1948, p. 77.

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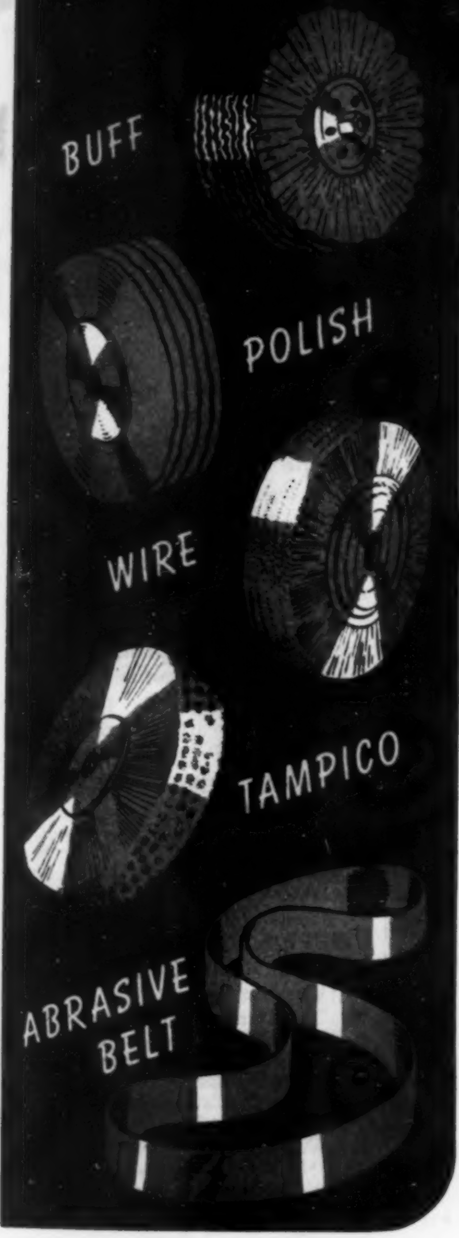

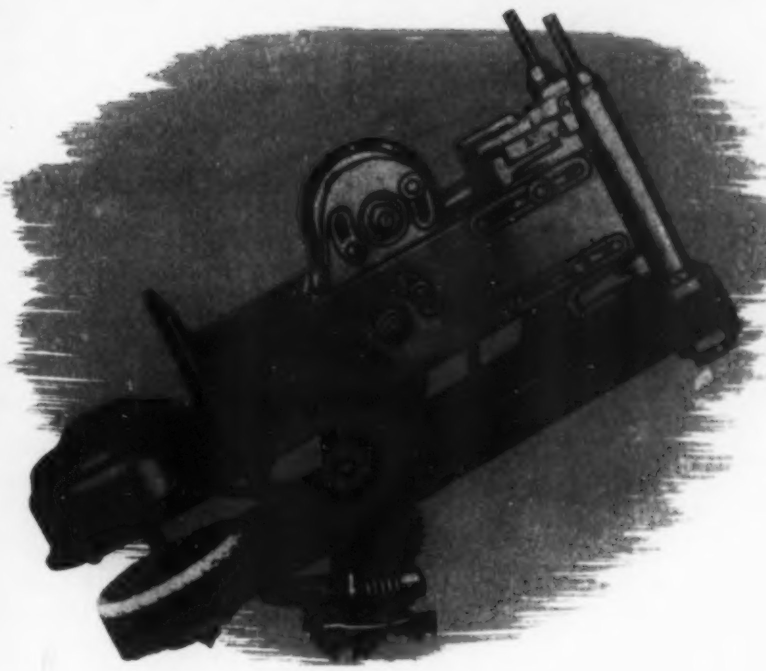

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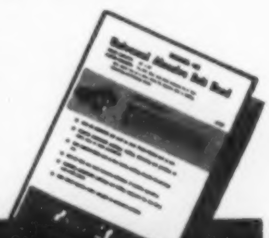

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Steel "Quality"

(Continued from p. 100)

He has noticed that British shipbuilders have had far less trouble than the American. He thought that the shipbuilding methods in both countries were sufficiently close to permit the assumption that the smoother going in England was not due to the differences in shipyard practices. He emphatically stated that in his opinion the only reason might be found in the quality of the British steel.

Brittle American Plate*

In a study of brittle fracture of American ship plates made by W. Barr of Messrs. Colvilles, Ltd., four sets of plates were examined which came from American ships that failed in service. Some were too badly corroded to permit examining the fractured edges, so samples were taken for micrographical, chemical and physical examination from localities close to and remote from the fracture.

Chemical analysis has shown all samples to be of good purity, only nitrogen being somewhat high. Samples taken from plates which apparently cracked worse than the others carried somewhat high carbon (0.26 and 0.29%) for mild steel, and in both samples the manganese was on the low side. Grain size was satisfactory and the microstructure normal except for a somewhat "rougher" pearlite in one.

Physical tests did not show any of the plates to be of low quality, but normalizing always brought about an enormous improvement in the Izod value and reduced the crystallinity of the fracture. In one steel a modest heat treatment for 1 hr. at 930° F. improved both the tensile strength and the Izod value. From these tests and other considerations the author concluded that strain-aging might be the fundamental cause of embrittlement.

A rimmed steel of low-carbon content has shown a very high Izod figure, but the fracture was strictly crystalline. Barr points out the extreme softness of that steel, which caused a large absorption of energy in bending, yet which does not mean that the sample was not brittle. *(Cont. on p. 104)*

**Engineering*, Feb. 27, 1948, p. 208.

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Brittle American Plate

(Continued from p. 102)

Further experiments compared American shipplates with those of British make. Eleven of the first and nine of the second were tested. The American steel averaged 0.23% carbon and 0.38% manganese; the British 0.18% carbon and 0.57% manganese. The latter were somewhat higher in tensile strength (which might be due also to the larger amount of scrap used, for the British analyzed over 0.10% of each of the tramp elements nickel, chromium and copper). As-rolled they have a higher Izod figure (23 against 17) and the average crystallinity of the fracture was about 80% as against 90% for American.

Structurally all steels were fairly uniform except for one American sample which exhibited fairly rough pearlite.

Normalizing improved both the American and British steels to a great extent. Izod results rose in the first to 46; in the second to 63. Annealing followed by furnace cooling at 150° per hr. lowered the Izod figures in both classes of steels, but the loss amounted to 55% for the American steels and to only 28% for the British.

Further tests covered the effects of temperature and of grain size upon the impact figures (Charpy) for American steel with 0.3 to 0.47% manganese and the British with 0.57 to 0.62%. All samples were previously normalized. American steels low in manganese had a transition point from partly to fully crystalline fracture somewhere around 60° F., while it was 30° F. for the high-manganese. At −4° F. the high-manganese British steel had still a modest impact strength, while the low-manganese American steel was quite brittle, and the one with 0.47% Mn just a little better.

Grain size had a great effect even in the tougher steels. Sizes 7 to 9 (normalized) had a transition point below −40° F., 6 to 8 about −30° F., and 1 to 4 around +40° F. The impact values at these temperatures were about 60.

An accelerated strain-aging test (10% strain and 480° F. for 30 min.) affected all samples badly. But fine-grained steels, so treated, had transition points at about 50° F. with an impact figure around 40, while the coarse-grained samples had a transition point around 120° F. At temperatures lower than 15° F. these differences lessened.